KAWASAKI STEEL TECHNICAL REPORT
No.4 (December 1981)
Special Issue on Steel Pipe

Development of Large Diameter High Strength Line Pipes for Low Temperature Services

Chiaki Shiga, Taneo Hatomura, Jun-ichi Kudoh, Akio Kamada, Keisuke Hirose, Toshihiro Sekine

Synopsis :
Large diameter high temperature line pipes of X65, X70, and X80 grades for the arctic gas transmission use have been developed by using controlled-rolled steel plates and by adopting a quench and temper treatment after forming the pipe. A combination of high strength and toughness without an increase in alloying elements can be provided by strictly controlled rolling of low C, low S, high Mn, and Nb- and V-bearing steel. Rolling in the dual phase ($\gamma + \alpha$) temperature range between Ar3 and "Ar3 - 40℃" can produce fine bainite and fine deformed ferrite grains in the ferrite-pearlite matrix, which is very effective in improving both strength and toughness of the plate and the pipe. By adopting the controlled rolling and the subsequent quench and temper treatment for low C and Mo- and Nb-bearing steel, BDWTT 85% shear FATT is significantly improved. This is due to the very fine microstructure of the plate which consists of fine ferrite and martensite island. The rolled steel plate of this type is proved to be useful for quenched and tempered pipes. The CVT, CV100 and BDWTT energy values of the pipes tested are much greater than those necessary for arresting unstable ductile fracture which have been shown in Battelle's full scale burst experiments.

The body can be viewed from the next page.

Development of Large Diameter High Strength Line Pipes for Low Temperature Services*

Chiaki SHIGA** Taneo HATOMURA** Jun-ichi KUDOH**
Akio KAMADA** Keisuke HIROSE*** Toshihiro SEKINE****

Large diameter high temperature line pipes of X65, X70, and X80 grades for the arctic gas transmission use have been developed by using controlled-rolled steel plates and by adopting a quench and temper treatment after forming the pipe.

A combination of high strength and toughness without an increase in alloying elements can be provided by strictly controlled rolling of low C, low S, high Mn, and Nb- and V-bearing steel. Rolling in the dual phase ($\gamma + \alpha$) temperature range between Ar_3 and "$Ar_3 - 40°C$" can produce fine bainite and fine deformed ferrite grains in the ferrite-pearlite matrix, which is very effective in improving both strength and toughness of the plate and the pipe.

By adopting the controlled rolling and the subsequent quench and temper treatment for low C and Mo- and Nb-bearing steel, BDWTT 85% shear FATT is significantly improved. This is due to the very fine microstructure of the plate which consists of fine ferrite and martensite island. The rolled steel plate of this type is proved to be useful for quenched and tempered pipes.

The CVT, CV100 and BDWTT energy values of the pipes tested are much greater than those necessary for arresting unstable ductile fracture which have been shown in Battelle's full scale burst experiments.

1 Introduction

Line pipes to be used in the arctic region not only require high strength and excellent toughness for preventing brittle fracture at both base metal and weld portions, but also must have great resistance to unstable ductile fracture peculiar to high pressure gas pipes. In addition, from the viewpoint of weldability in girth welding, it is required to reduce carbon equivalent to the minimum. Owing to the recent increase in gas pipeline transportation, the trends of line pipes are directed toward higher strength, thicker walls and larger diameters. However, it has been clarified by the results of the recent full scale burst tests[1-3] that pipes require higher impact absorbed energy value for arresting ductile fracture.

To meet such stringent requirements, Kawasaki Steel Corporation has been developing the following three manufacturing techniques: ① Controlled rolling method before pipe forming (CR)[4-6], ② A quench and temper treatment of the plate before pipe forming (QT), and ③ A quench and temper treatment of the pipe after its forming (pipe-QT). In the CR method, a low C and high Mn-Nb-V steel has been used as main material to develop a type of steel in which fine-grained bainite and fine-grained deformed ferrite have been introduced into its ordinary ferrite-pearlite matrix by effectively utilizing rolling at the dual phase ($\gamma + \alpha$) region, thereby achieving enhanced weldability and higher toughness due to decreases in alloying elements. To obtain pipe-QT steel, development has been under way for manufacturing low-cost steel for low temperature use that will replace conventional Ni steel, by applying Nb-Mo steel to the CR + QT process[7-9].

This report describes the results of laboratory experiments which have become the basis of the above developments, together with various mechanical properties of UOE pipes of the X65, X70, and X80 grades actually manufactured. The report also discusses that those pipes sufficiently satisfy the toughness values required for arresting unstable ductile fracture so far proposed.

* Originally published in *Kawasaki Steel Giho*, 13 (1981) 1, pp. 106-118
** Research Laboratories
*** Chiba Works
**** Mizushima Works

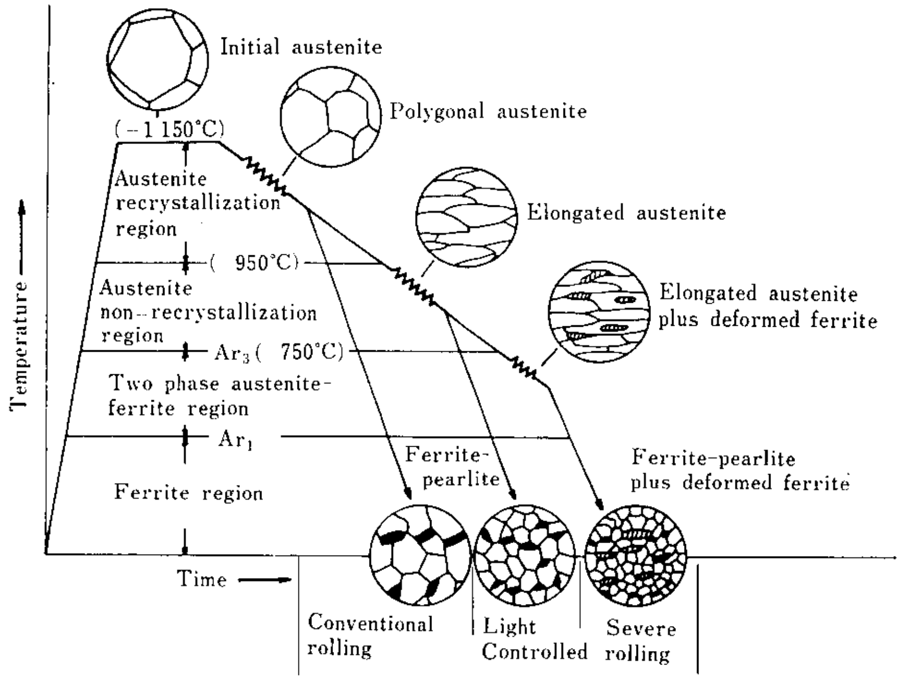


Fig. 1* Schematic illustration of controlled rolling process and resultant microstructure in comparison with conventional rolling for X70 niobium bearing steel

2 Controlled rolling

The temperature region for controlled rolling may be divided as shown in **Fig. 1** into three stages; ① Higher-temperature austenite recrystallization region, ② Austenite non-recrystallization region, and ③ Dual phase austenite-ferrite region. The figure indicates the schematic diagram of austenite structures in the process of rolling, when cumulative reduction up to those respective stages have been applied and final ferrite structures which have been obtained by cooling the austenite structures to room temperature. Compared with the ferrite in the conventional hot forming rolling in which the steel is finished at the higher-temperature recrystallization region exceeding about 900°C, the ferrite in the controlled rolling method becomes finer in grains as it comes down in rolling temperatures to the austenite non-recrystallization region and further down to the dual phase region. It is particularly noteworthy that fine-grained deformed ferrite is obtained by rolling at the dual phase region directly below the transformation temperature Ar_3.

According to the laboratory rolling experiments on Nb-bearing steel (M-1) in **Table 1**, an increase in the amount of reduction in the austenite non-recrystallization region makes, as is well known, the ferrite grains finer and improves the CVN 50% shear FATT as shown in **Fig. 2**. With rolling only in this region, however, there is a limit in making the grains finer, and as can be seen from **Fig. 3**, an increase in the amount of reduction in the preceding higher-temperature austenite recrystallization region is effective in improving the CVN 50% shear FATT. **Figs. 4** and **5** show mechanical properties and changes in separation for M-1 and M-2, respectively, when reductions of 62.5% and 50% are cumulatively applied at over 1 020°C and at 880 to 850°C, respectively, and 2 passes of a 30% reduction are applied for finishing at a temperature varying from 850 to 580°C. The transformation temperatures for M-1 and M-2 steels

Table 1 Chemical composition of the steels used

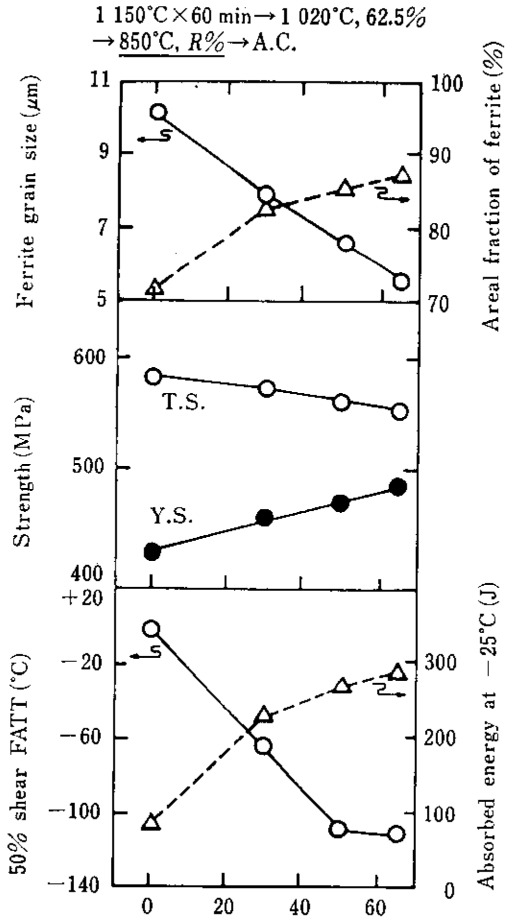


Fig. 2 Effect of the amount of reduction in austenite non-recrystallization region between 880°C and 850°C on the microstructure and properties in transverse direction of 16 mm thick plates for M-1 steel

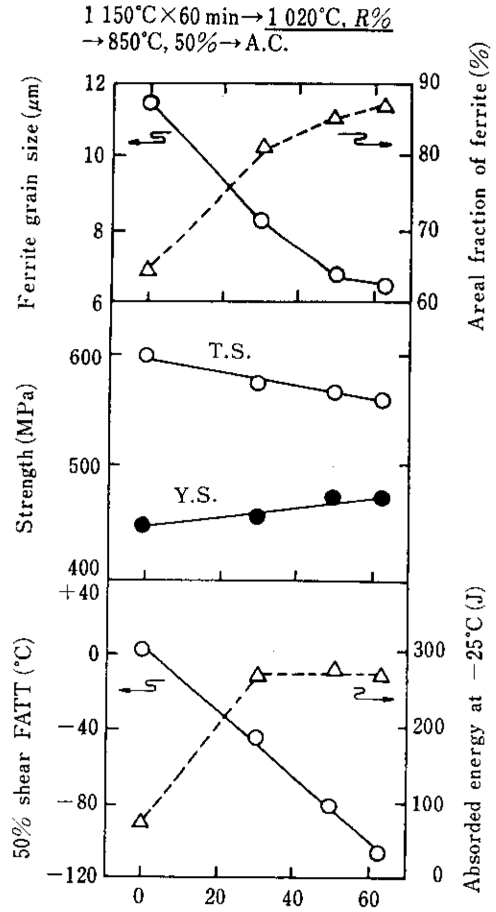
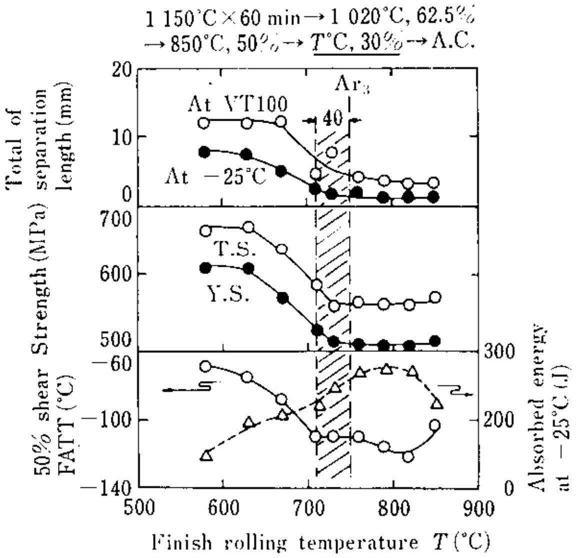


Fig. 3 Effect of the amount of reduction in austenite recrystallization region between 1 070°C and 1 020°C on the microstructure and properties in transverse direction of 16 mm thick plates for M-1 steel

are 755°C and 770°C, respectively. When the finishing temperature is lowered below the Ar_3 point, that is, when M-1 and M-2 steels are rolled at, say, 710°C and 730°C, respectively, the Y.S. and T.S. values rise sensitively, but the CVN 50% shear FATT remains practically the same. Only when the rolling temperature for M-1 steel drops to 670°C and that for M-2 steel drops to 710°C, will the CVN 50% shear FATT begin to rise. **Fig. 6** shows the relation between the amount of reduction in the temperature region of 730 to 710°C, properties and grain sizes of M-1 steel. An increase in the amount of reduction in this temperature region raises Y.S. and T.S., but is rarely accompanied with deterioration in the CVN 50% shear FATT. This may be attributable to the fact that ① Rolling in the very narrow dual phase region, i.e., in the width from Ar_3 to "Ar_3—40°C" in the hatched portion of **Fig. 5**, causes reduction strain to accumulate effectively in the austenite non-recrystallization grains, resulting in generation of fine-grained ferrite as shown in **Photo. 1**, and ② "Fine-grained deformed ferrite" which have been generated by deforming small grains that have already been transformed into ferrite but have not yet grown very much, increases tensile strength without adversely affecting toughness. When the finish-rolling temperatures for the respective steels are lowered further than "Ar_3— about 40°C," tensile strength increases while toughness deteriorates. The reason for this is that ferrite grains which have grown larger are rolled into larger deformed ferrite grains.

Rolling in the temperature region of "Ar_3— about 40°C" introduces fine-grained deformed ferrite and forms fine-grained ferrite, thereby making it possible to implement higher strength and higher toughness



VT100 : Lowest temperature of 100% shear fracture

Fig. 4 Effect of finish-rolling temperature on the separation and properties in transverse direction of 16 mm thick plates for M-1 steel

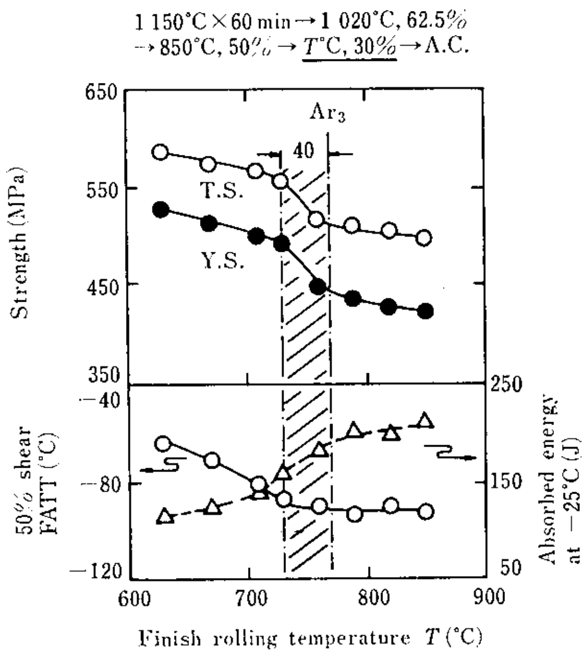


Fig. 5 Effect of finish-rolling temperature on the properties in transverse direction of 16 mm thick plates for M-2 steel

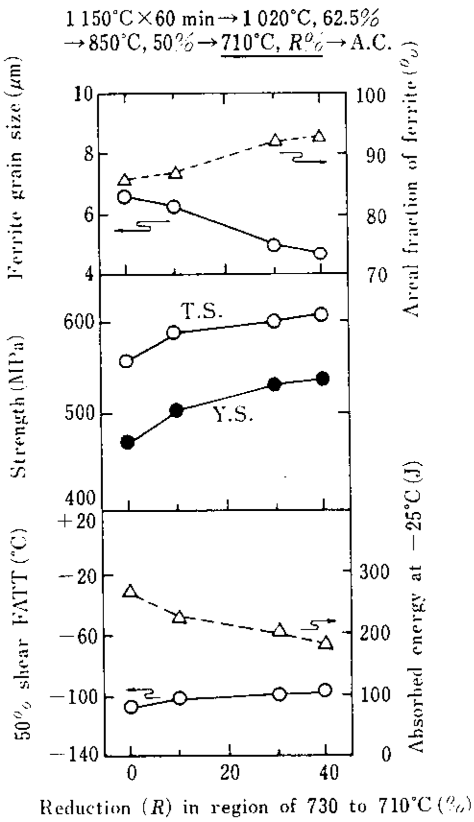
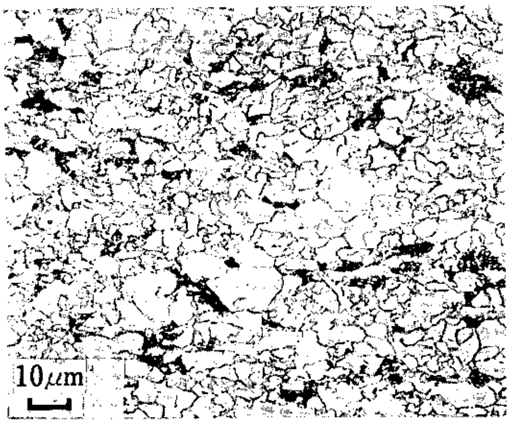


Fig. 6 Effect of the amount of reduction in dual phase region between 730°C and 710°C on the microstructure and properties in transverse direction of 16 mm thick plates for M-1 steel

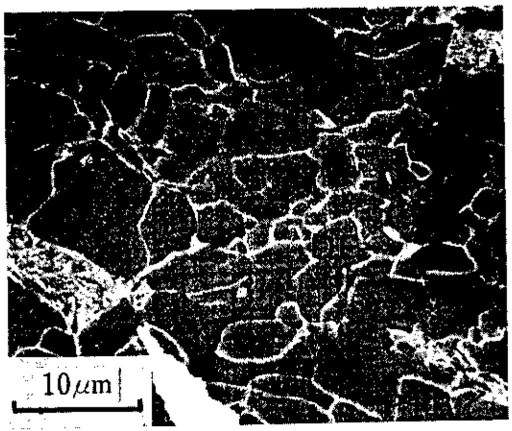
simultaneously with smaller alloying elements. This controlled rolling positively utilizing rolling in the narrow temperature region has been named "KTR (Kawasaki Thermomechanical Rolling)" by the authors. In performing rolling in this narrow dual-phase temperature region, it is important to achieve correct measurement of the plate temperature and correct detection of transformation temperature Ar_3. From the cooling curves in the laboratory rolling experiments, the Ar_3 points have been measured, and correlation between the Ar_3 point and the alloying elements such as C, Mn, Ni, Cr, Mo, and Cu, as expressed by the following equation has been obtained:

$$Ar_3(°C) = 910 - 273C\% - 74Mn\% - 56Ni\% - 16Cr\% - 9Mo\% - 5Cu\% \quad \cdots\cdots (1)$$

The Ar_3 point can also be obtained from the analysis using the surface temperature during rolling in the factory[6] and is used for actual rolling operation.



(a)



(b)

Photo. 1 Illustrative micrographs of M-1 steel plate finish rolled at 710°C, showing (a) general microstructure and (b) fine deformed ferrite (D.F.) and fine grained pearlite-bainite (P+B)

Fig. 7 shows the relation between the T.S. and the CVN 50% shear FATT when controlled-rolled by varying Ni and Mn contents in a Nb-bearing steel. Increases in Ni and Mn contents raise T.S. and, at the same time, lower the 50% shear FATT. One of the reasons for this is the lowering of the transformation temperature Ar_3 according to eq. **(1)** as a result of increases in these elements, because this lowering causes the austenite non-recrystallization region to expand to the lower side and suppresses grain growth after transformation, thereby making ferrite grains finer. Along with the increases in Ni and Mn contents, coarse-grained bainite becomes liable to form, but when rolling reduction in the austenite non-recrystallization region is sufficient, fine-grained bainite can be introduced into the ferrite-pearlite matrix. This fine-grained bainite contributes to both increased tensile strength and improved toughness.

Charpy impact absorbed energy is significantly affected by C and S contents. **Figs. 8** and **9** show the effects of C and S contents, respectively, on the Charpy V-notch absorbed energy in the transverse direction to rolling for steel plates of grades X60 to 70 controlled-rolled in the factory. A decrease in C content reduces the amounts of pearlite and bainite, resulting in enhancing absorbed energy. A decrease in S content reduces the amount of elongated MnS, thereby greatly enhancing absorbed energy of the base material. In addition, as shown in **Fig. 10**, the decrease in S content contributes to the enhancement of HAZ toughness and anti-hydrogen induced cracking properties.

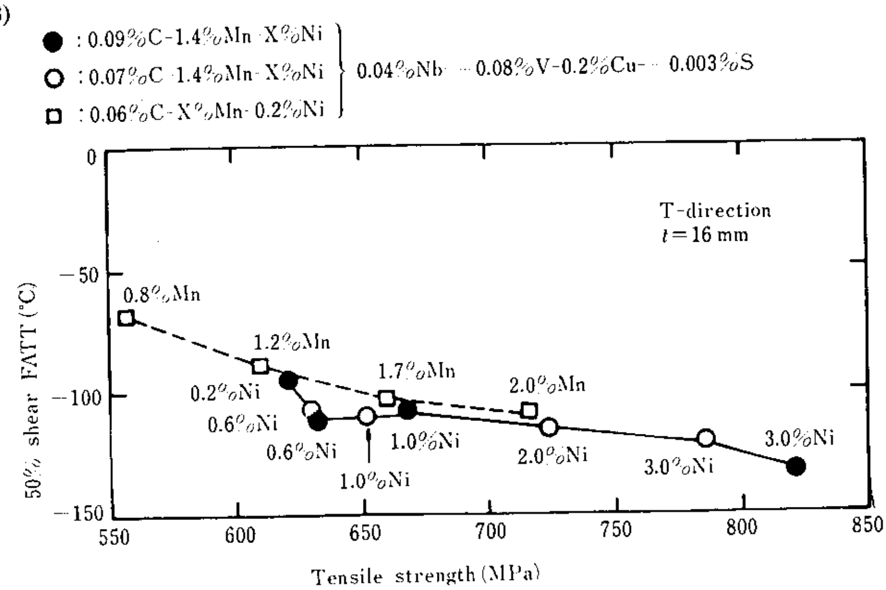


Fig. 7 Relation between Charpy V-notch transition temperature (50% FATT) and tensile strength of controlled-rolled steels with varied manganese or nickel content

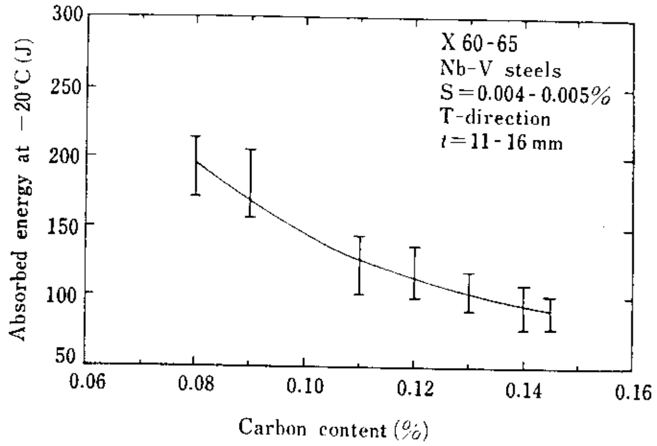


Fig. 8 Effect of carbon content on the Charpy V-notch absorbed energy of controlled-rolled niobiumvanadium steel plates for X60 to X65 pipes manufactured in the factory

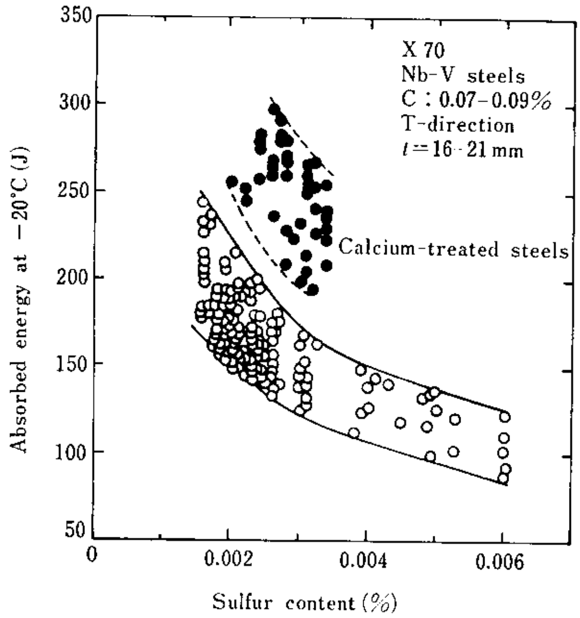


Fig. 9 Effect of sulfur content on the Charpy V-notch absorbed energy of controlled-rolled niobium-vanadium steel plates for X70 pipes manufactured in the factory

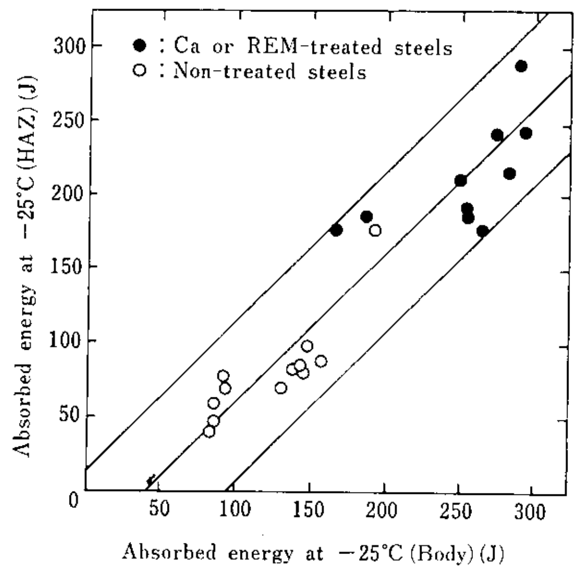


Fig. 10 Effect of Ca or REM treatment on Charpy V-notch absorbed energy at −25°C in the body and heat affected zone (HAZ) of approx. 20 mm thick UOE pipes

3 Development of Steel for QT Pipe Use

In the manufacturing method in which the pipe, after its forming, is subjected to induction heating and the quench and temper treatment, higher tensile strength can be obtained at a lower carbon equivalent than in the pipe which has been subjected to only controlled rolling, and furthermore higher impact absorbed energy can be more easily obtained. Another major feature is that the toughness of the HAZ and Bond in the pipe weld portion is raised to the same degree as the toughness of the base material. Since this manufacturing step is a process of increasing the tensile strength of the pipe after its forming, the tensile strength at the time of pipe forming can be made lower, and thus the manufacture of high tensile pipes above the X80 grade becomes easier, viewed from the pipe making capacity. However, it is generally difficult to obtain the BDWTT 85% shear FATT at a temperature below −25°C and the CVN 50% shear FATT at a temperature below −100°C for the quenched- and tempered-plates or pipes whose steels contain less than 1% of Ni.

Fig. 11 shows the effect of the amount of rolling

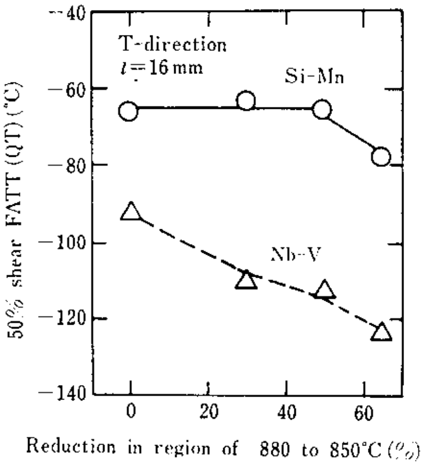


Fig. 11 Effect of the amount of reduction in austenite non-recrystallization region between 880°C and 850°C on the CVN 50% shear FATT of Si-Mn (M-3) and Nb-V (M-1) steels after quench and temper treatment

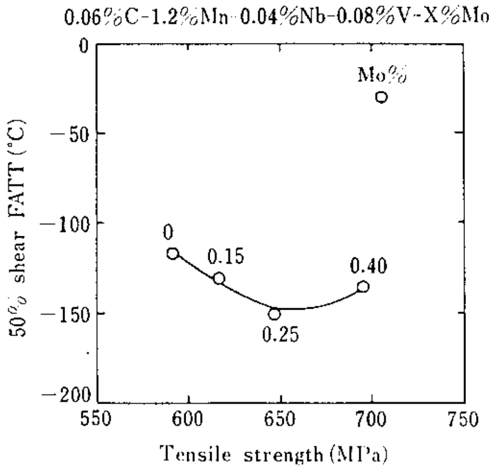


Fig. 12 Effect of molybdenum content of CVN 50% shear FATT and tensile strength of Nb steel plates produced by CR+QT process

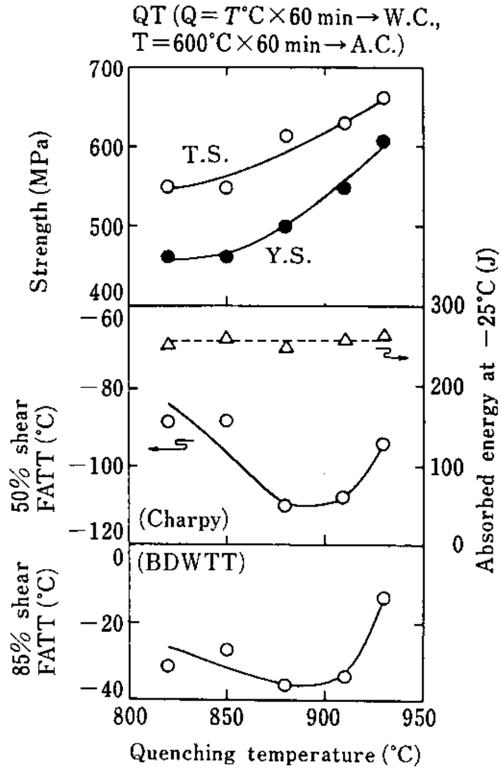


Fig. 13 Effect of quenching temperature on the mechanical properties in transverse direction of 19 mm thick Nb-Mo steel plates produced by CR+QT process

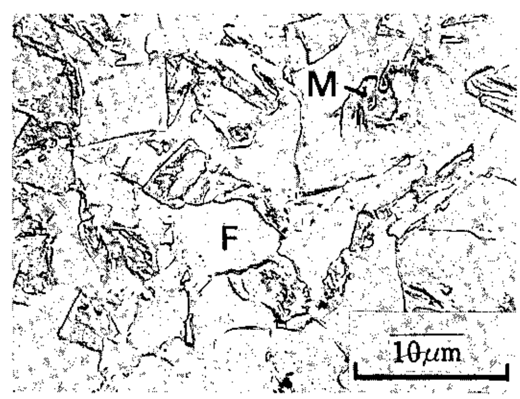


Photo. 2 Electron replica micrographs showing fine ferrite grains (F) and martensite island (M)

reduction in the austenite non-recrystallization region on toughness after QT of Nb-bearing steel (M-1) and Si-Mn steel (M-3). In the Nb-bearing steel, contrary to the Si-Mn steel, the CVN 50% shear FATT after QT becomes below −100°C by a 50% reduction in the austenite non-recrystallization region. Namely, in the Nb-bearing steel, the reduction hysteresis before QT significantly influences toughness after QT. The reason for this is that in proportion to the fineness of the as-rolled ferrite grains, ferrite grain sizes after QT become smaller and their occupancy ratio becomes larger. The effect of the reduction hysteresis becomes more conspicuous when rapid heating such as induction heating is applied. Namely, by using the process of "CR + induction QT," it is possible to obtain a super-fine structure mainly consisting of ferrite.

Fig. 12 shows the effect of Mo contents on the CVN 50% shear FATT and tensile strength of Nb-steel plates produced by the "CR + QT" process. An increase in Mo content increases T.S. and lowers the

CVN 50% shear FATT simultaneously. This is attributable to the fact that martensite island is added to fine ferrite grains and the so-called dual phase structure has been formed as shown in **Photo. 2. Fig. 13** shows the relation between the quenching temperature and the properties of 19 mm thick Nb-Mo steel plates produced by the "CR + QT" process. The quenching temperature that gives satisfactory CVN 50% shear FATT and BDWTT 85% shear FATT lies in the temperature region between the point directly under the transformation temperature Ac_3 and about 30°C.

4 Line pipes for Use in Arctic Region

On the basis of the fundamental facts mentioned above, gas line pipes for arctic region use were manufactured by the UOE method. **Tables 2** and **3** show chemical compositions and mechanical properties, respectively, of line pipes of the X65, X70 and X80 grades which measure 48″ and 56″ in outside diameter and 0.6–1.4″ in wall thickness. In **Table 3**, these pipes are broadly classified into the following three:

(1) Pipes whose service temperature is −25°C and which have high impact absorbed energy
(2) Pipes whose service temperature is −25°C and which have very high impact absorbed energy
(3) Thick-walled pipes whose service temperature is −60°C and which have high impact absorbed energy

These three kinds of pipes have high impact absorbed energy which is enough to arrest unstable ductile fracture. Pipes of (1) are abbreviated to "HE pipes (high energy pipes)" and pipes of (2) to "EHE pipes (extra high energy pipes)".

The EHE pipes have a Charpy impact absorbed energy of more than 230 J at −25°C or a CV100 energy of more than 136 J (100 ft · lbf).

The steel plates for the HE pipes were manufactured by controlled-rolling of low C-low S-high Mn-Nb steel, to which addition contents of Ni and Mo were increased as the wall thickness increased. The steel plates for the EHE pipes upto the grade X70 were manufactured by controlled rolling of a Ca-treated low C-low S-high Mn-Nb-V steel, and the EHE pipes of grade X80 was manufactured by quenching-tempering of pipe from a Nb-Mo steel.

Those CR pipe steels feature the low C-low S-high Mn contents, and designs of their contents were based on the results of **Figs. 7–9**. In controlled-rolling process, all the steels were slab-soaked at about 1 150°C and reduced at a ratio of about 60% in the higher temperature austenite recrystallization region, at a ratio of about 70% in the austenite non-recrystallization region, and at a ratio of about 25% in the dual phase region. In all cases pipe steel plates were rolled from thick slabs. The HE pipe steel plates were finish-rolled at a temperature below "Ar_3 −40°C," whereas the EHE pipe steel plates were finish-rolled in the temperature region of Ar_3 to "Ar_3 −40°C."

The grade X80 pipe was made by applying the pipe-QT treatment to the UOE pipe with an induction heater, whose mother plate was controlled-rolled in the austenite non-recrystallization region with about 70% reduction. Quenching was performed by water jetting to both the inner and outer surfaces of the pipe. As a result, the steel with a carbon equivalent of 0.38% gave the strength of the X80 grade and a BDWTT 85% shear FATT of −35°C; and as shown in **Figs. 14** and **15**, the weld portion did not develop hardened and

Table 2 Chemical composition of line pipe steels for the Arctic use

(%)

Steel	C	Si	Mn	P	S	Ni	Mo	Cu	Cr	Nb	V	Al	Ca	REM	C_{eq}*
C 1	0.06	0.27	1.68	0.019	0.005	0.20		0.20	—	0.04	0.03	0.036	—	—	0.373
C 2	0.05	0.29	1.74	0.016	0.004	—			0.18	0.05	—	0.040	—	—	0.376
C 3	0.06	0.26	1.70	0.014	0.005		0.14		—	0.05	—	0.036	—	—	0.371
C 4	0.06	0.27	1.63	0.019	0.005	1.10	—	—	—	0.05	0.04	0.032	—	—	0.413
C 5	0.05	0.26	1.70	0.018	0.002	0.21	—	0.24	—	0.04	0.03	0.033	Add.	—	0.369
C 6	0.06	0.21	1.69	0.007	0.003	0.30	—	—	—	0.04	0.03	0.030	—	Add.	0.368
C 7	0.07	0.25	1.62	0.023	0.003	0.21	—	0.24	—	0.04	0.07	0.030	Add.		0.384
Q 1	0.07	0.25	1.35	0.004	0.003	—	0.18	—	0.19	0.03	0.04	0.034	—		0.377
Q 2	0.05	0.16	1.44	0.006	0.004	2.77	0.21	—	0.22	—	—	0.031	—	—	0.561
Q 3	0.07	0.25	1.06	0.013	0.011	3.45	0.19	—	0.15	—	0.02	0.039	—	—	0.549

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

Table 3 Mechanical properties of line pipes for the Arctic use

(1) Grade X65 and X70 pipes with high Charpy absorbed energy for service temperature of −25°C (High Energy Pipe)

Grade	Steel	Pipe size, wall thickness outside diameter mm (inch)	Production process	Body										Weld portion	
				Tensile test				Charpy V notch test				DWTT		Charpy V notch test	
														HAZ	Weld
				Y.S.	T.S.	Y.R.	El.	at −25°C	CV100*	VT100**	50% FATT	at −25°C	85% shear	at 25°C	
				MPa		%		J		°C		J	°C	J	
X70	C 1	21.3×1 219 (0.84×48)	CR	511	592	86.2	43	126	88	−80	−115	7 490	−35	96	165
X70	C 2	〃	〃	512	603	84.8	41	138	98	−80	−120	—	−48	91	71
X65	C 1	26.9×1 219 (1.06×48)	〃	468	576	81.2	43	129	74	−100	−135	—	−42	95	172
X65	C 3	〃	〃	499	583	85.7	47	156	105	−80	−110	—	−44	64	188
X65	C 4	32.5×1 219 (1.28×48)	〃	475	590	80.5	42	125	83	−80	−115	11 700	−60	75	163

* CV100 is energy at VT100, **VT100 is lowest temperature of 100% shear fracture

(2) Grade X70 and X80 pipes with extra high Charpy absorbed energy for service temperature of 25°C (Extra High Energy Pipe)

Grade	Steel	Pipe size, wall thickness outside diameter mm (inch)	Production process	Body											Weld portion			
				Tensile test				Charpy V notch impact test					DWTT		Charpy V notch impact test			
															HAZ		Weld	
				Y.S.	T.S.	Y.R.	El.	at −25°C	at −80°C	CV111	CV100	50% FATT	at 25°C	85% shear	at −25°C	at 40°C	at 25°C	at 40°C
				MPa		%		J			°C		J	°C	J			
X70	C 5	15.2×1 219 (0.60×48)	CR	509	605	84.1	39	261	140	177	−69	−112	5 886	−53	247	226	152	140
X70	C 6	18.3×1 219 (0.72×48)	〃	492	587	83.8	42	293	137	139	−70	−115	7 151	−63	213	188	155	144
X70	C 7	25.4×1 219 (1 ×48)	〃	502	603	80.0	50	290	209	209	−80	−120	13 538	−45	192	187	164	174
X80	Q 1	19.1×1 422 (0.75×56)	Pipe QT*	586	683	86.0	40	235	203	235	−40	−105	10 301	−35	203	196**	186	167**

* Induction heating, ** at −80°C

(3) Grade X70 heavy walled pipes for service temperature of −60°C

Grade	Steel	Pipe size, wall thickness outside diameter mm (inch)	Production process	Body									Weld portion	
				Tenisle test				Charpy V notch test			DWTT		Charpy V notch test	
													HAZ	Weld
				Y.S.	T.S.	Y.R.	El.	at 60°C	at 100°C	50% FATT	85% shear		at 60°C	
				MPa		%		J		°C	°C		J	
X70	Q 2	25.4×1 219 (1 ×48)	QQT*	500	634	77.1	50	220	183	−136	−65		79	124
X70	Q 3	35.6×1 219 (1.4×48)	QQT*	508	659	77.1	46	140	108	<−140	−74		65	98

* Duplicate quenching and tempering, roller quench and furnace heating

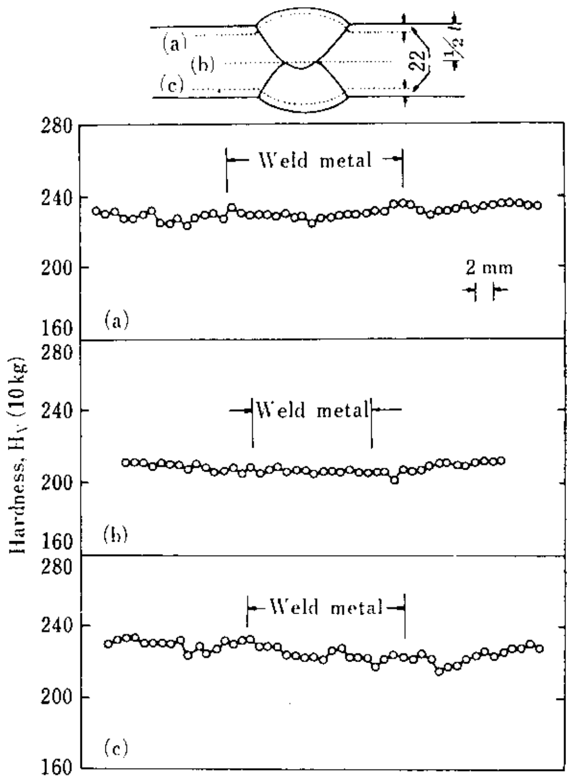


Fig. 14 Hardness distribution of seam welded portion in grade X80 pipe manufactured by PIPE-QT process

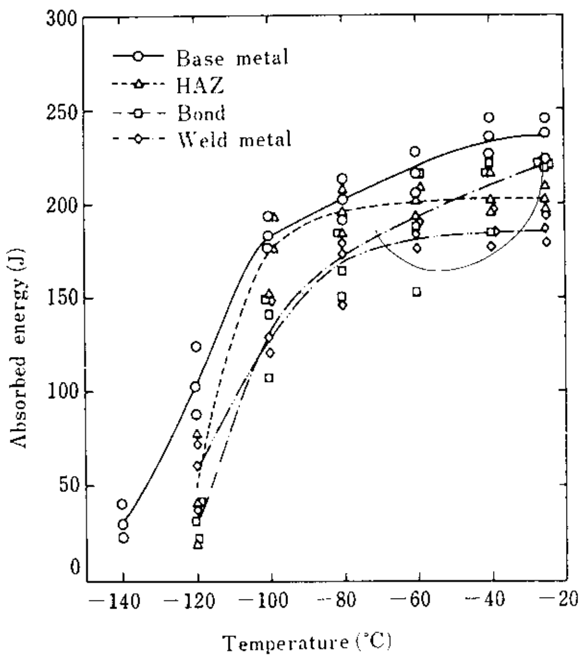


Fig. 15 Charpy impact absorbed energy curves of body and welded portion in grade X80 pipe manufactured by PIPE-QT process

softened zones and the fusion line of heat-affected zone gave a CVN 50% shear FATT of −80°C.

Pipes of (3) are thick-walled with the service temperature of −60°C. These steels have been used by Kawasaki Steel for sometime as low temperature service steel plates. They contain 2.8 to 3.5% Ni and have been made by applying UOE pipe forming to a steel plate which was given the duplicate quench and temper treatment.

5 Properties for Arresting Propagation of Unstable Ductile Fracture

Large diameter steel pipes to be used as high pressure gas linepipes must have sufficient resistance to unstable ductile fracture. As a measure for expressing resistivity to unstable ductile fracture, Charpy impact absorbed energy is widely used. On the basis of the results of many full-scale burst tests so far conducted, Charpy energy values (CVN) necessary for arresting unstable ductile fracture were proposed by Battelle's Columbus Laboratories (B.C.L.)[1], American Iron and Steel Institute[10] and British Gas Corporation[11]. Among these values, the highest energy value is the one proposed by B.C.L., which gives eq. **(2)**, depending upon pipe shapes and pressure values:

$$CVN = 3.57 \times 10^{-5}(\sigma_H)^2(Rt)^{0.33} \quad \cdots\cdot(2)$$

CVN: 2 mm V full size Charpy energy (J)
σ_H : Hoop stress (MPa), $\sigma_H = PR/t$
P : Internal pressure of pipe (MPa)
R : Radius of pipe (mm)
t : Wall thickness of pipe (mm)

The eq. **(2)** is a simplified equation obtained for the soil-backfilled pipe. For the non-backfilled pipe, it is necessary to return to the B.C.L. theory[1] and obtain CVN necessary for the ductile fracture arrest from the conditions that the so-called J-curve comes into tangent contact with the intra-pipe decompression curve. According to the calculation result, CVN necessary for the arrest in case of a non-backfilled pipe is about 1.2 times the CVN value for the backfilled pipe obtained from eq. **(2)**. As CVN value, CVT or CV100 is used where CVT is the energy value at the service temperature of pipes, and CV100, the energy value at the lowest temperature to exhibit 100% shear area. However, CV100 has ambiguity in its definition and physical meaning, and further in the recent B.C.L. studies[12], it is reported that the use of this value would tend to make the required energy for ductile fracture arrest overestimated.

G.M. Wilkowski et al.[12], on the other hand, reported that particularly for steel plates exhibiting separation, the "effective Charpy energy (CVN)," which is calculated from the following eq. **(3)** by using the BDWTT energy (DTE (J)) at the service temperature of the pipe, can express the resistance value of material against unstable ductile fracture better than the Charpy impact absorbed energy:

$$(DTE/A_D) = 0.631 + 3\,(CVN/A_C) \cdots\cdots (3)$$

A_D: Net sectional area of BDWTT specimen (mm^2)

A_C: Net sectional area of full-size Charpy specimen (mm^2)

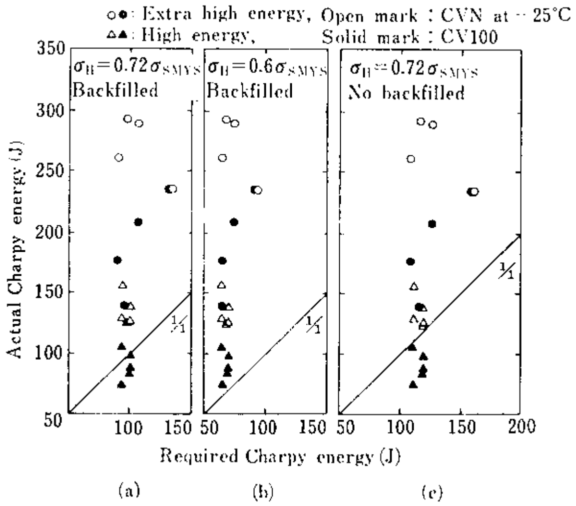


Fig. 16 Comparison of actual Charpy energy of the pipes shown in Table 3 and Charpy energy required for ductile fracture-arrest under the following conditions; (a) $\sigma_H = 0.72\sigma_{SMYS}$, backfilled, (b) $\sigma_H = 0.6\sigma_{SMYS}$, backfilled and (c) $\sigma_H = 0.72\sigma_{SMYS}$, no-backfilled

Fig. 16(a), (b), and (c) show a comparison between the CVN values required for ductile fracture arrest obtained from eq. **(2)** and the actual CVT and CV100 values of the respective pipes in respect of HE and EHE pipes in **Table 3**. **Fig. 17**(a), (b), and (c) show a comparison between the CVN values, which have been calculated by eq. **(2)** and converted into BDWTT energy DTE by using eq. **(3)**, and the actual BDWTT energy values of the respective pipes. In **Figs. 16** and **17**, it is assumed that the service temperature of the soil-back-filled peipeline is −25°C and operating pressure values at hoop stress are $\sigma_H = 0.60\,\sigma_{SMYS}$ and $\sigma_H = 0.72\,\sigma_{SMYS}$ (σ_{SMYS}: specified minimum yield stress). Further for $\sigma_H = 0.72\,\sigma_{SMYS}$, a comparison has been made between the required and actual values of impact absorbed energy in case of the no backfilled peipeline. **Figs. 16** and **17** clearly indicate that for HE pipes shown with the △ and ▲ marks the use of any one of CVT, CV100 and DTE as a measure of fracture toughness satisfies the required value under $\sigma_H = 0.60\,\sigma_{SMYS}$, whereas under $\sigma_H = 0.72\,\sigma_{SMYS}$, whether they are soil-backfilled or non-backfilled, only CV100 does not satisfy the required value.

For the EHE pipes indicated with ○ and ● marks, the actual impact energy value of CVT, CV100 and DTE significantly exceeds the required value under any operating conditions.

As mentioned above, CV100, which has ambiguity in its definition and physical meaning, is considered to be a required value which is overly on the conservative side. Therefore even if the actual CV100 value is not satisfied with the required value, the pipe is considered

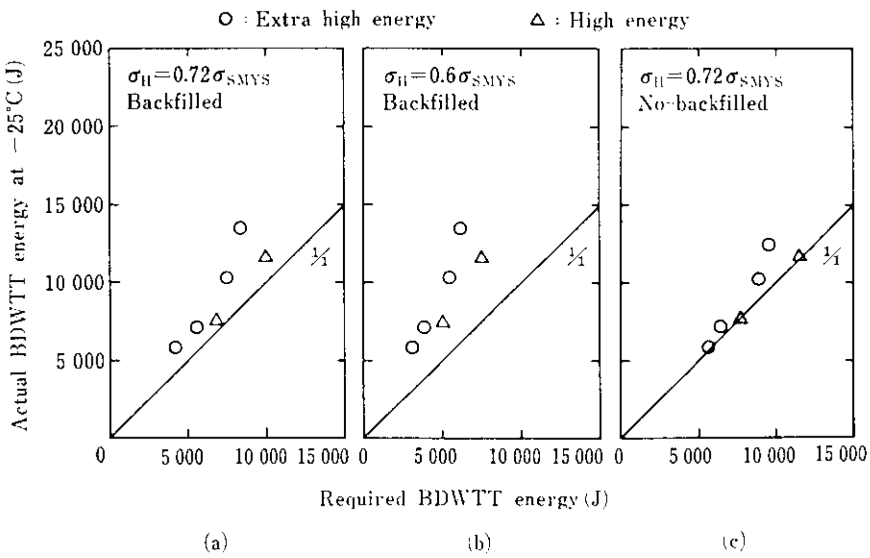


Fig. 17 Comparison of actual BDWTT energy of the pipes shown in Table 3 and BDWTT energy required for ductile fracture-arrest under the following conditions; (a) $\sigma_H = 0.72\sigma_{SMYS}$, backfilled, (b) $\sigma_H = 0.60\sigma_{SMYS}$, backfilled and (c) $\sigma_H = 0.72\sigma_{SMYS}$, no-backfilled

to have sufficient resistance to unstable ductile fracture if CVT and DTE satisfy the required energy. Namely, HE pipes are considered to have sufficient ductile fracture arrestability under normal operation pressure of $\sigma_H = 0.6$ to $0.72\,\sigma_{SMYS}$. The EHE pipes have room for ductile fracture arrestability even under $\sigma_H = 0.72\,\sigma_{SMYS}$, and are considered to be able to play satisfactorily the role of the so-called "crack arrester pipe" even under more severe operating conditions. For instance, suppose that economy is given priority in pipeline construction, and a greater part of the pipeline is to be constructed with low-cost, low-impact energy pipes, but in several places at certain intervals, high-impact energy pipes (crack arrester pipes) which can sufficiently arrest unstable ductile fracture are to be installed. In such case the EHE pipes will be optimal pipes.

6 Conclusion

Laboratory experiments were conducted on the controlled rolling (CR) for plates and quenching and tempering (QT) of pipes with an induction heater in order to develop high-strength large-diameter line pipes for arctic service. On the basis of the above, X65 to X80 grades UOE pipes with outside diameters of 48″ and 56″ were manufactured. Conclusions are summarized below:

(1) The temperature region of CR is divided into the following three stages: Higher temperature austenite recrystallization region, austenite non-recrystallization region and dual phase austenite-ferrite region. Rolling between Ar_3 and "Ar_3 − 40°C" can raise tensile strength without sacrificing toughness. This is attributable to the formation of "fine-grained deformed ferrite."

(2) In the relation between the Ar_3 point of steel during rolling and chemical composition, the following equation becomes valid:

$$\begin{aligned} Ar_3(°C) = 910 &- 273C\% - 74Mn\% \\ &- 56Ni\% - 16Cr\% \\ &- 9Mo\% - 5Cu\% \end{aligned}$$

As a result the above rolling in the dual phase region can be completely controlled.

(3) Charpy impact absorbed energy in the direction transverse to the rolling direction of CR steel plate rises exponentially in proportion to decreases in C and S contents. Further, with Ca or REM treatment, the absorbed energy value rises much higher.

(4) The Ca or REM treatment of CR steel improves the toughness of both base material and heat affected zone.

(5) Increases in Ni and Mn contents of CR steel raise tensile strength and simultaneously lower the impact fracture appearance transition temperature. This is attributable to fine-grained bainite that is grown in place of pearlite, besides fine-grained ferrite.

(6) When Nb-bearing steel is controlled-rolled, reheated at a temperature slightly lower than Ar_3 point and then quenched and tempered (CR + QT), its toughness is significantly improved. This is attributable to the fact that the structure of quenched-and-tempered steel containing niobium succeeds the rolling hysteresis. Namely, fine ferrite grains in as-rolled plate contribute to make finer the austenite grains in reheating and furthermore the ferritic and bainitic structure after subsequent quenching.

(7) When Nb-Mo steel is given CR + QT, it has the X80 grade in tensile strength and an excellent toughness in BDWTT which exceeds those of conventional QT steel. This is attributable to the formation of a super-fine-grained structure consisting of fine-grained ferrite and martensite island.

(8) The following three types were manufactured: ① Pipes of X65 and X70 grades with a service temperature of −25°C, a high Charpy impact absorbed energy of about 130 J, and a wall thickness of 0.84 to 1.28″ (HE pipes), ② Pipes of X70 and X80 grades with a service temperature of −25°C, a very high absorbed energy of over 230 J, and a wall thickness of 0.6 to 1″ (EHE pipes) and ③ Grade X70 pipes with a service temperature of −60°C and a wall thickness of 1 to 1.4″.

(9) Charpy impact absorbed energy values of HE and EHE pipes at −25°C satisfy energy values necessary for arresting unstable ductile fracture set forth in the proposal by Battelle's Columbus Laboratories which is considered the severest.

(10) The EHE pipes have impact energy well enough to arrest unstable ductile fracture. When they are evaluated by any of the energy values of CV100, CVT and BDWTT, namely, they can satisfactorily play the role of the "crack arrester pipe."

References

1) W. A. Maxey et al.: Symposium on "Crack Propagation in Pipelines," Inst. Gas. Eng. (London), (1974) (paper 16)
2) W. A. Maxey and R. J. Eiber: "Materials Engineering in the Arctic," *Proceedings of ASM Conference*, (1976), pp. 306–319
3) L.S.I.J. High Strength Line Pipe Research Committee, Symposium on Pipe, API, (1980)
4) T. Tanaka, T. Funakoshi, M. Ueda, J. Tsuboi, T. Yasuda and C. Utahashi: *Proc. of Microalloying* '75, (1975), Union Carbide Corp., N.Y., p. 38

5) T. Tanaka, N. Tabata, T. Hatomura and C. Shiga: *Proc. of Microalloying* '75, (1975), Union Carbide Corp., N.Y., p. 350
6) Y. Saito, N. Koshizuka, C. Shiga, T. Sekine, T. Yoshizato and T. Enami: *Proc. of "Science and Technology of Flat Rolled Products"*; ISIJ, JAPAN, September, (1980)
7) T. Hatomura, C. Shiga, A. Kamada and N. Ohashi: *Trans. ISIJ*, **20** (1980), B96
8) C. Shiga, T. Hatomura, A. Kamada and N. Ohashi: *Tetsu-to-Hagané*, **65** (1970), S487
9) C. Shiga, H. Ohtsubo, A. Kamada, N. Ohashi, K. Hirose and H. Mottate: International Conference—Pipeline and Energy Plant Piping—"*Fabrication in the '80s*," Calgary, Canada, November, (1980) pp. 10–13
10) AISI Committee of Large Diameter Line Pipe Producers: "Running Shear Fractures in Line Pipe," (1974), AISI, New York
11) W. A. Poynton: Symposium on "Crack Propagation in Pipelines", Inst. Gas. Eng. (London) (1974)
12) G. M. Wilkowski: 6th Symposium on Line Pipe Research, AGA, (1979)