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TENSILE PROPERTIES OF LOWER CARBON TRIP STEELS

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ABSTRACT

The uniaxial tensile properties of a series of TRIP steels of varying carbon contents and processing histories were determined over a wide range of test temperatures.

The yield strengths at room temperature varied both with the deformation temperature (over the range 250° to 550°C) and with the carbon content (0.05 to 0.20%). Possible reasons for these variations were advanced.

For all steels, the yield strength exhibited a minimum at a test temperature of about -50°C and a maximum at a test temperature of about 100°C. The minima and the maxima were especially pronounced for the steels processed at the lowest deformation temperatures.

Both the rate of work hardening and the elongation were influenced by the strain-induced austenite-to-martensite transformation. Both the rate of strain hardening and the rate of production of strain-induced martensite (per unit strain) increased with decreasing temperature.

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Tensile Properties of Lower Carbon TRIP Steels

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INTRODUCTION

The introduction of high-carbon high-manganese steels by Sir Robert Hadfield eighty-six years ago inaugurated the use of strain-induced phase transformations to improve the mechanical properties of steels.⁽¹⁾ Since that time, this phenomenon has been widely exploited, particularly for enhancing the properties of metastable austenitic stainless steel.⁽²⁻⁵⁾ Both Hadfield's manganese steel and metastable austenitic stainless steels have low yield strengths and high elongations in the solution-quenched condition, and in the cold-worked condition they have high yield strengths and low elongations. Both of these combinations of strength and ductility are useful, but it would be better to have high strength combined with high elongation. In a recent paper, Zackay et al.⁽⁶⁾ described a process for producing high-strength steels with high values of elongation. These steels were designed to be thermodynamically unstable so that plastic straining would induce a martensitic transformation. A suggestion was made in this paper that steels exhibiting a high degree of transformation induced plasticity be called TRIP steels.

The present study is concerned with the effects of the warm working used to raise the yield strength and the testing temperature on the tensile properties of several lower carbon TRIP steels. The carbon contents of these steels were chosen to bracket the estimated equilibrium eutectoid carbon content of a base alloy containing 12 Cr, 8 Ni, 4 Mo, 1.5 Si, 0.75 Mn. In this study, the primary emphasis was placed on

correlating the properties of the steels with the structure. A wide range of processing and testing conditions was explored.

EXPERIMENTAL PROCEDURE AND RESULTS

Twenty pound vacuum melted ingots were forged to 0.5 in. thickness at 1100°C and subsequently hot rolled to 0.25 in. at 900°C. The plates were solution annealed at 1100°C for one hour and water quenched. Final deformations during warm working (which involved 80% reduction in thickness except where noted) were carried out at temperatures of 250°, 350°, 450°, and 550°C. Preheated rolls were used and temperature control was maintained by returning the pieces to the furnace between passes to re-establish temperature equilibrium. The compositions of the four steels investigated are given in Table I.

Sheet tensile specimens having a one inch gauge length, a thickness of 0.05 in., and a test section width of 0.125 in., were ground from processed sheets. The specimens were loaded by means of pins passing through holes to the enlarged end sections to avoid misalignment. The total elongation was measured between small indentations made on the surface prior to testing. A yield point occurred in most cases (except for the solution-quenched steels), and the yield stress was taken as the stress where the load dropped. When there was no drop in load, the 0.1% offset method was used to obtain the yield stress. True stress-strain calculations were based on measurements of engineering stress-strain data taken from the Instron recorder. (The total elongation as measured at the end of the test was used as the scaling factor.) The elastic strain of both the specimen and the tensile machine was subtracted from the total strain in computing these curves. The strain rate employed was 0.04 per

TABLE I

Steel Designations and Percentages of Alloying Elements

Steel Designation	C	Cr	Ni	Mo	Si	Mn	Fe
A	0.05	12.1	7.7	3.9	1.5	1.1	Bal.
B	0.07	12.1	7.8	3.9	1.5	0.83	Bal.
C	0.16	12.1	7.8	3.9	1.5	0.82	Bal.
D	0.20	12.0	7.9	4.0	1.5	0.80	Bal.

minute; for test temperatures above and below room temperature, the specimen was immersed in a temperature controlled liquid.

The amount of the transformation that occurred during testing was determined quantitatively by a magnetization method. The technique was based upon measurements of the difference in saturation magnetization of austenite and martensite. The readings were converted to volume percent martensite, and corrections were made for the influence of the alloying elements. (7,8) Measurements of the volume fraction of martensite produced during straining at room and liquid nitrogen temperatures were made during tensile testing. For other testing temperatures, measurements were made only at room temperature before and after the test.

Steels A, B, C, and D had different M_s temperatures because of differences in carbon content. The M_s temperatures of steels A and B were above 22°C ; both contained some martensite when the material was quenched to room temperature, as shown in Figs. 1a and 1b. These two steels were stabilized against further decomposition by room temperature aging. After holding at 22°C for several weeks, no additional martensite formed in steel A until it was cooled to -35°C ; and steel B did not transform further until it was cooled below -70°C . Steels C and D contained less than 0.1 volume percent martensite even after cooling to -196°C .

The test data are summarized in Tables II through VI. The qualitative magnetic responses (i.e. to a hand magnet) of the specimens before and after tensile testing are also included in these tables. The true stress-true strain curves of two steels (0.05% and 0.20%C) with prior deformation at 250°C are shown in Figs. 2 and 3 for

several test temperatures. (These curves are plotted to the point of maximum load, not to fracture.)

DISCUSSION

Deformation Temperature

The deformation temperature is an important processing variable in the production of TRIP steels because it influences both the strength and the stability of the austenite.⁽⁹⁾ In an earlier paper, the effect of the amount of deformation on the tensile properties of these steels was discussed.⁽⁶⁾ In the present study, the amount of deformation was held constant (80%), and the deformation temperature was varied from 250° to 550°C. The mechanical working can be done at any temperature above M_D (M_D is the temperature above which no martensite will form during plastic deformation.) The influence of deformation temperature on the room temperature yield strengths and elongations of steels A, B, C, and D, is shown in Figs. 4 and 5.

The yield strengths of the steels of intermediate carbon content (0.07% and 0.16%) are not significantly influenced by the deformation temperature, as is shown in Fig. 4. However, at the 250°C deformation temperature, the yield strength of the 0.05 carbon steel is low, while that of the 0.20% carbon steel is high. The reasons for these differences in behavior will be considered next.

The yield strength of the 0.05% carbon steel is relatively low for prior deformation at 250° and 350°C. It rises to a slight maximum for 450°C, and, finally, decreases again to a lower value after processing at 550°C. The low yield strength for the 250° and 350°C deformation temperatures is believed to be due to the martensite (9%) which was contained in the steel after the solution treatment and which became tempered during the rolling. Tempered low carbon martensite is weak, and

its presence is believed to be responsible for the lower yield strength. The tempered martensite can be seen in Fig. 6, which is a photomicrograph of the specimens deformed at 250°C. The somewhat higher strength of the 0.05% carbon steel after deformation at 450°C can be attributed to the presence of additional untempered martensite that formed during cooling from the deformation temperature. Measurements showed that the total amount of martensite had increased from the original 9 to 22 volume percent after the 450°C processing treatment.

After processing at 550°C, the amount of martensite was only 14 volume percent, and the yield strength was lower than after the 450°C treatment. In this case, it is thought that more carbon was retained in solution because of the higher solubility of carbon in austenite at the higher temperature and that this carbon retention made the austenite more stable.

A distinctive feature of the plots shown in Fig. 4 is the high yield strength of the 0.20% carbon steel deformed at 250°C. The higher yield strength is presumed to be due to hardening caused by carbide precipitation during the 250°C processing treatment. Metallographic observations support the carbide precipitation hypothesis. The undeformed, solution-treated materials contained some undissolved carbides, the amount being greater in the higher carbon steels, as can be seen by a comparison of the photomicrographs of solution quenched steels A and B in Fig. 1 and steel D in Fig. 7. After deformation, the amount of precipitation was observably greater in all of the steels, as is shown by comparisons of Fig. 1 with Fig. 6 and Fig. 7 with Fig. 8. Furthermore, the amount of

precipitate formed during deformation was greatest for the steel of the highest carbon content, as can be seen by a comparison of Fig. 6 with Fig. 8.

In most instances, the yield strength and the ultimate tensile strength increased with carbon content for each deformation temperature studied, as is shown in Tables III through V. However, the values for steel B were sometimes lower than for those of steel A. This was presumably due to the larger quantity of martensite initially present in steel A. The deformation temperature does not appear to influence the elongation significantly, as was shown in Fig. 5.

Consistent with the proposed role of carbon in TRIP steels are the properties of steels of the same nominal composition (9 Cr, 8 Ni, 4 Mo, 2 Mn, 2 Si) and similar processing (80% deformation at 450°C) but differing in carbon content⁽⁶⁾ (0.25% and 0.30%). The yield strength and elongation for the 0.25% C steel were 208,000 psi and 27% and those for the 0.30% C steel were 222,000 psi and 31%. The expected greater volume fraction of precipitated carbides in the 0.30% C steel evidently accounts for the higher yield strength observed with the higher carbon alloys.

Test Temperature

The temperature dependence of the tensile properties of TRIP steels is complex. The stability of the austenite as well as the flow characteristics of both austenite and strain-induced martensite are influenced by the temperature of testing.⁽¹⁰⁻¹⁴⁾ The variations of yield strength and elongation with test temperature for steels A, B, C, and D, in both the solution quenched and deformed (80% at 250°C) conditions are shown in Figs. 9 and 10. The yield strengths of the solution quenched steels increase monotonically with decreasing temperature at all of the test temperatures employed.

For all steels of the series, regardless of the deformation temperature, the yield strength increased with decreasing test temperature between -50° and -196°C. The increase in yield strength for the deformed steels was greater than for those that were undeformed in the -50° to -196°C range. In general, the deformed steels exhibited a minimum in yield at about -50°C and a maximum at about 100°C. The minima and maxima were especially pronounced for the steels deformed at 250°C. Similar trends were observed for the temperature dependence of the yield strength of solution quenched AISI Type 304 stainless steel.⁽¹⁵⁾

A striking feature of TRIP steels is the sharp drop in elongation above the M_D temperature (which is estimated for these steels to be between 100° and 200°C). Above the M_D , the austenite no longer transforms to martensite during straining; and the elongation approaches that of highly cold-worked austenite. As several investigators have shown,^(16,17) the formation of martensite during straining enhances the work hardening

of metastable austenitic steels and necking of tensile test specimens is thereby inhibited.

The ultimate tensile strengths of all four steels in both the solution quenched and deformed conditions exhibited the strong temperature dependence that is characteristic of metastable austenitic steels⁽¹⁸⁾ (see Figs. 11 and 12). This behavior can be attributed to the high work-hardening rates resulting from the formation of strain-induced martensite.

The existence of a wide range of work hardening rates in TRIP steels is revealed by true-stress true-strain curves made at several test temperatures for the 0.05 and 0.20% C steels deformed 80% at 250°C (shown in Figs. 2 and 3). As indicated by the dashed lines, all specimens tested below the M_D deform initially (for 5 to 10% strain) by the formation and growth of Luders' bands. Following the formation of the Luders' bands throughout the entire gauge length, the steels work-hardened rapidly, with the rate of work-hardening increasing with decreasing test temperature. For the special case of the steels tested at 100°C, the near-horizontal curve suggests that the entire strain was due to Luders' band formation and growth.

The influence of test temperature on the rate of work hardening of both the 0.05% and 0.20% steels, as solution quenched and as deformed, is shown in Fig. 13. The rate of work hardening was determined by measuring the slope of the true-stress true-strain curve at a true strain slightly beyond the Luders' strain.

The work-hardening rate increased markedly with decreasing test temperature below the M_D , as shown in Fig. 13. This is reflected in the increase in the amount of martensite produced per unit strain, as shown

in Fig. 14. The rate of work hardening of these steels was further enhanced by prior deformation. At all test temperatures, the deformed, and hence stronger, steels had a higher rate of work hardening than the solution-quenched steels.⁽¹⁹⁾ Apparently, the kind of martensite produced in a deformed austenite matrix is more effective in hardening than that formed in solution-quenched austenite. This may be due to a finer plate size and/or a higher defect density in the martensite.⁽²⁰⁾

The work-hardening rate for both the deformed and solution-quenched steels becomes very low as the amount of martensite produced during straining approaches zero, (i.e., at temperatures above about 100°C), as can be seen in Fig. 13. The low work-hardening rate, however, was not as detrimental to the elongation of the solution quenched steels, (see Fig. 10), because their lower strength did not require a high rate of work hardening to prevent necking.⁽²¹⁾ The deformed steels quickly necked and failed at low elongations when they were tested at temperatures at or above the M_D , as shown in Fig. 10. Below the M_D , the rate of work hardening of the deformed steels is dependent upon the amount of strain-induced martensite produced per unit strain. The correlation between work-hardening rate and the rate of martensite formation is evident from the plots shown in Figs. 13 and 14. This behavior is consistent with the observations of Gunter and Reed,⁽¹⁵⁾ Bannerjee, et al,⁽¹⁶⁾ Bressanelli and Moskowitz,⁽¹⁷⁾ and Cina,⁽²²⁾ among others.

Influence of Alloying Elements

The alloy content of the steels studied to date is such that the amount of carbon that can be retained in solid solution is severely restricted. As previously discussed, the eutectoid carbon content for these steels is approximately 0.15%. As is well known, strong carbide formers such as molybdenum, vanadium, titanium, and niobium severely depress the carbon solubility in austenitic steels. (23) Zackay et al. (6) assumed in an earlier investigation that a fairly large amount of a strong carbide former such as molybdenum was necessary for optimum properties. In a recent investigation, Fahr (24) has shown that this is not necessary. He has obtained results which indicated that TRIP steels free of strong carbide-forming elements can be made with excellent combinations of strength and elongation. For example, the yield strength and elongation of a fully austenitic TRIP steel, (nominal composition 9 Cr, 8 Ni, 2 Mn, 0.42 C), deformed 80% at 250°C, were 248,000 psi and 34% respectively. This result suggests that TRIP steels of higher carbon content can be made with an accompanying improvement over the properties obtained to date.

SUMMARY

The uniaxial tensile properties of a series of TRIP steels of varying carbon content and processing histories have been determined over a wide range of test temperatures. The results can be summarized as follows:

The yield strengths at room temperature were independent of deformation temperature from 250° to 550°C for the steels of intermediate carbon content (0.07 and 0.16%). However, the yield strength of the lowest carbon steel (0.05%) was low and that of the highest carbon steel (0.20%) was high for the lowest deformation temperature (250°C). Possible reasons for this behavior were advanced, viz. decomposition of the austenite during processing for the low carbon steel and precipitation hardening coupled with work hardening for the high carbon steel.

The ultimate tensile strengths and the elongations at room temperature were relatively insensitive to the deformation temperature for all the steels of the series.

For all steels of the series, regardless of deformation temperature, the yield strength exhibited a minimum at a test temperature of about -50°C and a maximum at a test temperature of about 100°C. The minima and maxima were most pronounced for steels deformed at 250°C.

The ultimate tensile strengths of all four steels in both the solution quenched and deformed conditions also exhibited a strong dependence on test temperature.

The rate of work hardening and the elongation were influenced by the strain induced transformation, especially in the deformed steels. Above

the M_D temperature, both the rates of work hardening and the elongations of the deformed steels were low.

Well below the M_D temperature, the rates of work hardening and the elongations of the deformed steels were high, reflecting the presence of strain-induced martensite. The rate of production of strain-induced martensite per unit strain paralleled that of the rate of strain hardening in that both increased with decreasing temperature. At any temperature below M_D the amount of martensite produced per unit strain was greatest for the lowest carbon (the least stable) steel.

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TABLE II
Properties of Steels as Solution Quenched*

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elongation*** %	-Magnetic Response****	
					Before Test	After Test
-196	0.05	64	289	23	M	M
	0.07	48	309	20	M'	M
	0.16	78	248	24	A	M
	0.20	87	257	24	A	M
-78	0.05	52	195	30	M'	M
	0.07	39	213	26	A	M
	0.16	61	213	44	A	M
	0.20	68	222	48	A	M
22	0.05	42	140	38	M'	M
	0.07	37	158	30	A	M
	0.16	50	141	64	A	M
	0.20	48	133	59	A	M
100	0.05	32	83	49	M'	M
	0.07	28	79	40	A	M
	0.16	35	90	66	A	M'
	0.20	37	94	62	A	M'
200	0.05	25	74	40	M'	M'
	0.07	22	70	44	A	A

TABLE II cont.

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
200	0.16	28	78	46	A	A
	0.20	28	80	43	A	A

* Austenitized at 1100°C for one hour, water quenched.

** The 0.1% offset method used for determining the yield strength except for the 0.05 and 0.07 percent carbon steels tested at -196°C.

*** 1" gauge length, thickness 0.050", width 0.125".

**** M = magnetic, M' = weakly magnetic, A = non-magnetic or slightly magnetic.

TABLE III

Properties of Steels After 80% Deformation at 250°C*

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elongation*** %	Magnetic Response****	
					Before Test	After Test
					-196	0.05
	0.07	186	322	18	M'	M
	0.16	191	330	18	A	M
	0.20	209	346	19	A	M
- 78	0.05	136	235	16	M'	M
	0.07	142	251	19	M'	M
	0.16	163	261	20	A	M
	0.20	170	267	21	A	M
22	0.05	131	192	17	M'	M
	0.07	148	206	19	M'	M
	0.16	161	214	22	A	M
	0.20	191	227	26	A	M
100	0.05	168	170	23	M'	M
	0.07	193	193	23	M'	M
	0.16	190	190	23	A	M
	0.20	209	209	17	A	M
200	0.05	162	174	4	M'	M'
	0.07	162	175	5	M'	M'

TABLE III cont.

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
200	0.16	169	179	5	A	A
(cont.)	0.20	174	186	4	A	A

* Austenitized at 1100°C for one hour and water quenched prior to deforming at 250°C.

** The 0.1% offset method used for determining the yield strength for the four tests at 200°C.

*** 1" gauge length, thickness 0.050", width 0.125".

**** M = magnetic, M' = weakly magnetic A = non-magnetic or slightly magnetic.

TABLE IV

Properties of Steels After 80% Deformation at 350°C*

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before	After
					Test	Test
-196	0.05	177	327	20	M	M
	0.07	174	344	22	M'	M
	0.16	179	348	19	A	M
	0.20	189	328	14	A	M
- 78	0.05	125	219	17	M	M
	0.07	139	241	18	M'	M
	0.16	147	250	20	A	M
	0.20	145	259	15	A	M
22	0.05	128	177	18	M'	M
	0.07	144	205	20	A	M
	0.16	149	213	19	A	M
	0.20	156	215	20	A	M
100	0.05	150	152	23	M'	M
	0.07	167	175	35	A	M
	0.16	174	179	34	A	M
	0.20	182	190	29	A	M
200	0.05	148	158	4	M'	M'
	0.07	157	166	5	A	A

TABLE IV. cont.

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
200	0.16	168	175	5	A	A
(cont.)	0.20	173	183	5	A	A

* Austenitized at 1100°C for one hour and water quenched prior to deforming at 350°C.

** The 0.1% offset method used for determining the yield strength for the four tests at 200°C.

*** 1" gauge length, thickness 0.050", width 0.125".

**** M = magnetic, M' = weakly magnetic, A = non-magnetic or slightly magnetic.

TABLE V

Properties of Alloys After 80% Deformation at 450°C*

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
-196	0.05	187	235	12	M	M
	0.07	169	322	23	M	M
	0.16	174	348	24	A	M
	0.20	188	361	25	A	M
-78	0.05	155	215	17	M	M
	0.07	142	236	20	A	M
	0.16	142	248	21	A	M
	0.20	146	254	19	A	M
22	0.05	153	185	23	M	M
	0.07	150	199	22	A	M
	0.16	153	210	24	A	M
	0.20	161	218	23	A	M
100	0.05	168	168	5	M	M
	0.07	160	163	28	A	M
	0.16	181	183	28	A	M
	0.20	177	182	36	A	M
200	0.05	163	171	3	M	M
	0.07	155	162	4	A	A

TABLE V cont.

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T,S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
200	0.16	162	169	3	A	A
(cont.)	0.20	163	171	5	A	A

* Austenitized at 1100°C for one hour and water quenched prior to deforming at 450°C.

** The 0.1% offset method used for determining the yield strength for the four tests at 200°C.

*** 1" gauge length, thickness 0.050", width 0.125".

**** M = magnetic, M' = weakly magnetic, A = non-magnetic or slightly magnetic.

TABLE VI

Properties of Alloys Alloys After 80% Deformation at 550°C*

Test Temp. °C	Carbon Content %	Y.S.** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
-196	0.05	170	226	15	M	M
	0.07	176	279	15	M'	M
	0.16	175	337	22	A	M
	0.20	182	259	13	A	M
- 78	0.05	149	225	16	M	M
	0.07	131	238	18	A	M
	0.16	140	253	20	A	M
	0.20	144	251	20	A	M
22	0.05	145	191	19	M	M
	0.07	136	207	20	A	M
	0.16	146	216	19	A	M
	0.20	163	225	23	A	M
100	0.05	157	157	11	M	M
	0.07	157	174	27	A	M
	0.16	151	176	35	A	M
	0.20	178	189	30	A	M
200	0.05	131	141	3	M	M
	0.07	130	148	6	A	A

TABLE VI cont.

Test Temp. °C	Carbon Content %	Y.S,** (1000 psi)	T.S. (1000 psi)	Elonga- tion*** %	Magnetic Response****	
					Before Test	After Test
200	0.16	146	157	4	A	A
(cont.)	0.20	147	163	4	A	A

* Austenitized at 1100°C for one hour and water quenched prior to deforming at 550°C.

** The 0.1% offset method used for determining the yield strength for the four tests at 200°C.

*** 1" gauge length, thickness 0.050", width 0.125".

**** M = magnetic, M' = weakly magnetic, A = non-magnetic or slightly magnetic.

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FIGURE CAPTIONS

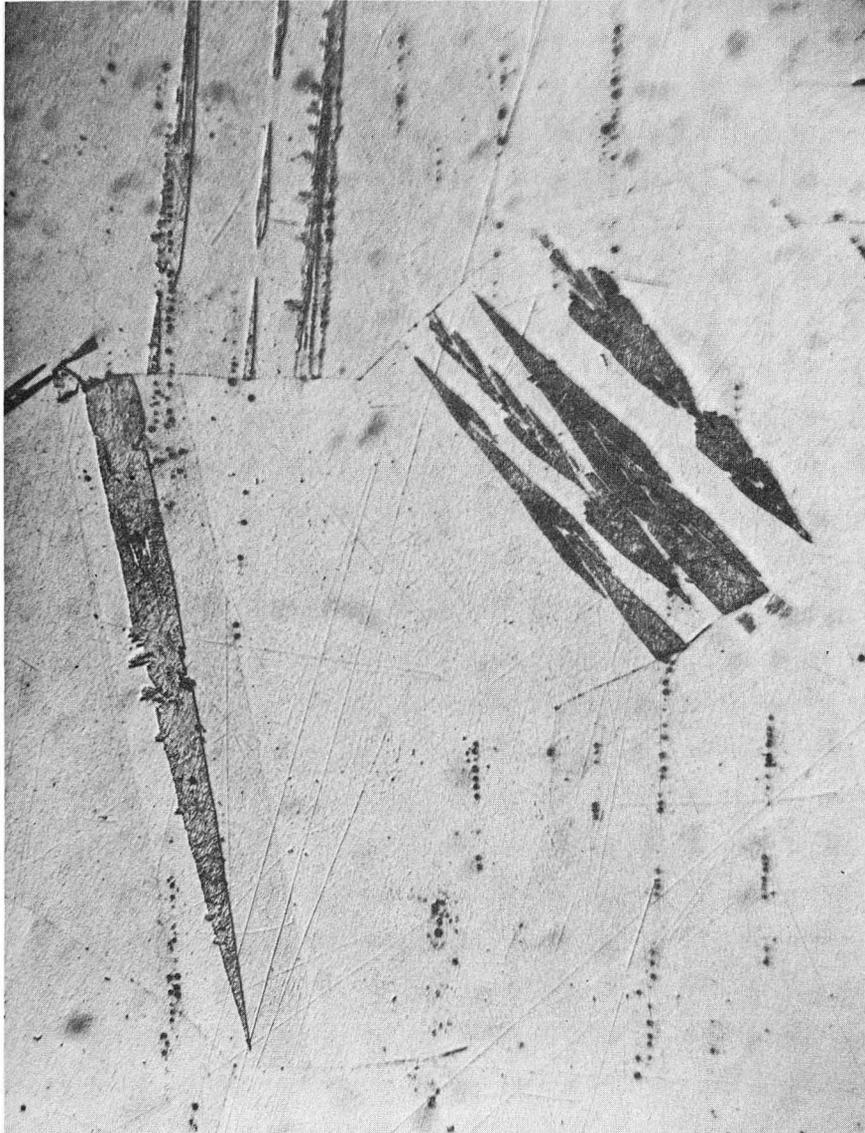
- Fig. 1 The microstructures of steels A and B showing the presence of (a)-(b) martensite. Both were solution annealed and quenched to room temperature: (a) steel A, (b) steel B. Magnification 900X.
- Fig. 2 True-stress true-strain curves for steel A (deformed 80% at 250°C) tested at several temperatures.
- Fig. 3 True-stress true-strain curves for steel D (deformed 80% at 250°C) tested at several temperatures.
- Fig. 4 The room temperature yield strengths of steels A, B, C, and D, for several deformation temperatures.
- Fig. 5 Room temperature elongations of steels A, B, C, and D, for several deformation temperatures.
- Fig. 6 The microstructure of steel A as deformed 80% at 250°C. Dark streaks are regions containing tempered martensite. Magnification 900X.
- Fig. 7 The microstructure of steel D as solution annealed and quenched to room temperature. Magnification 900X.
- Fig. 8 The microstructure of steel D as deformed 80% at 250°C. Magnification 900X.
- Fig. 9 The yield strengths of both deformed and solution-quenched steels at several test temperatures.
- Fig. 10 Elongations of both deformed and solution-quenched steels at several test temperatures.
- Fig. 11 The ultimate tensile strengths of the solution-quenched steels at several test temperatures.

FIGURE CAPTIONS cont.

Fig. 12 The ultimate tensile strengths of the deformed steels at several test temperatures.

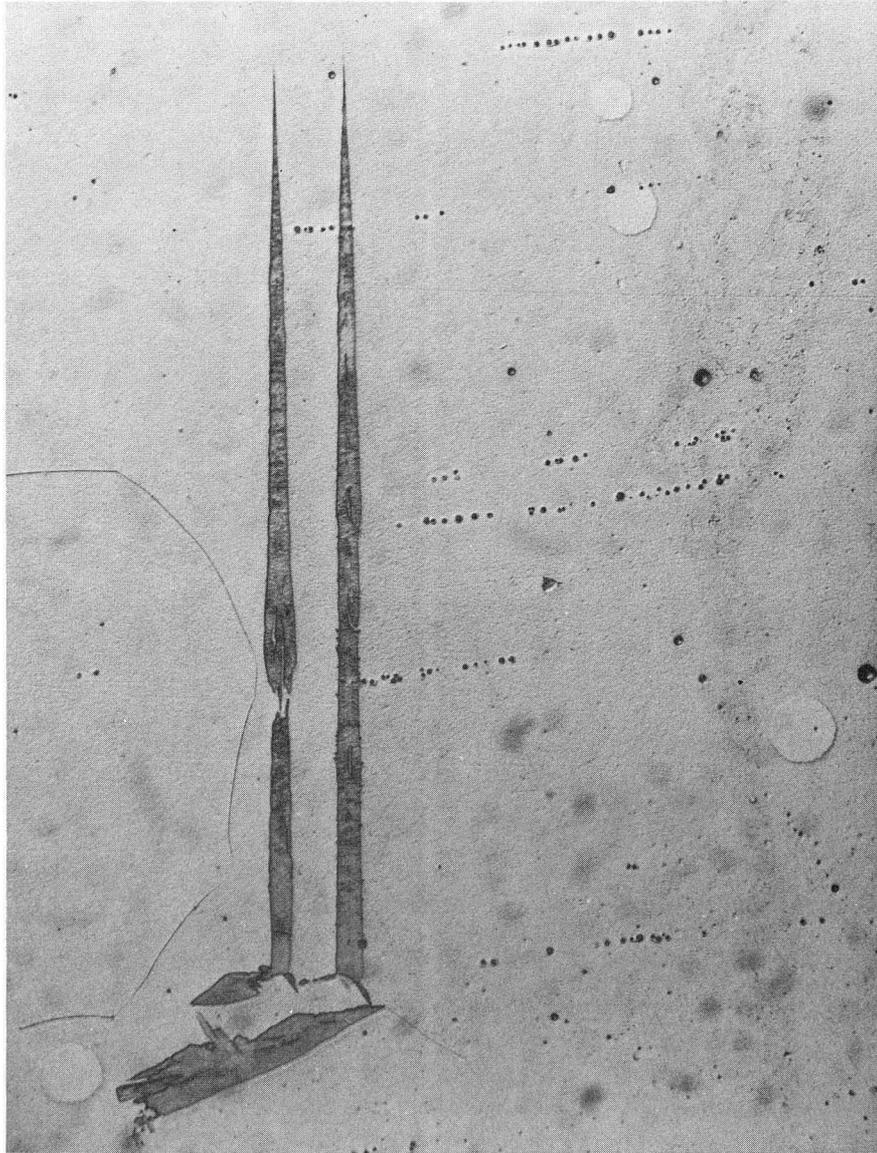
Fig. 13 The rates of work hardening of solution-quenched and of deformed A (0.05C) and D (0.07C) steels at several test temperatures.

Fig. 14 The rate of martensite production per unit strain for steel D in the deformed condition. (80% at 250°C.)



XBB 697-4647

Fig. 1(a)



XBB 697-4646

Fig. 1(b)

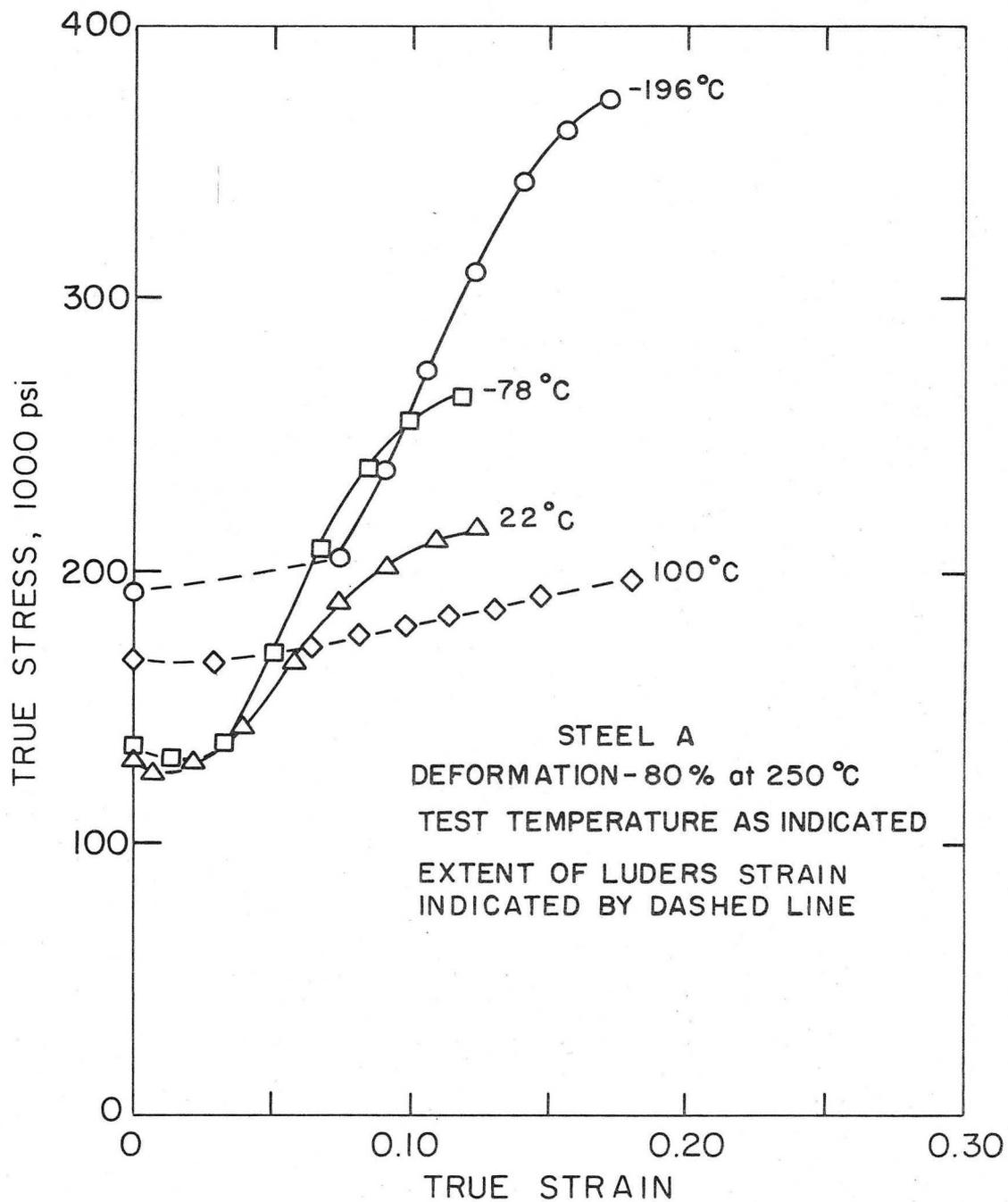


Fig. 2

XBL 689-5874

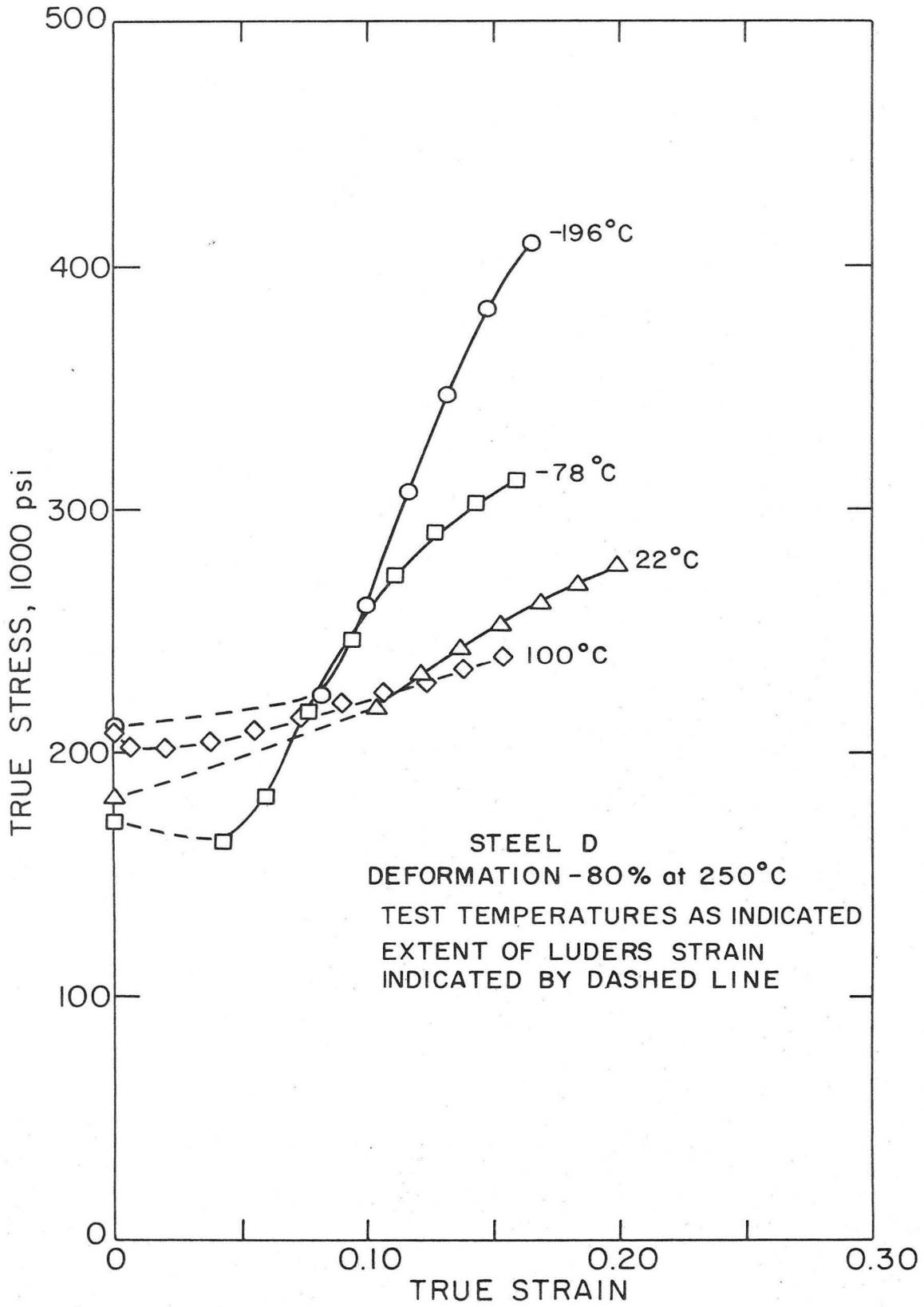
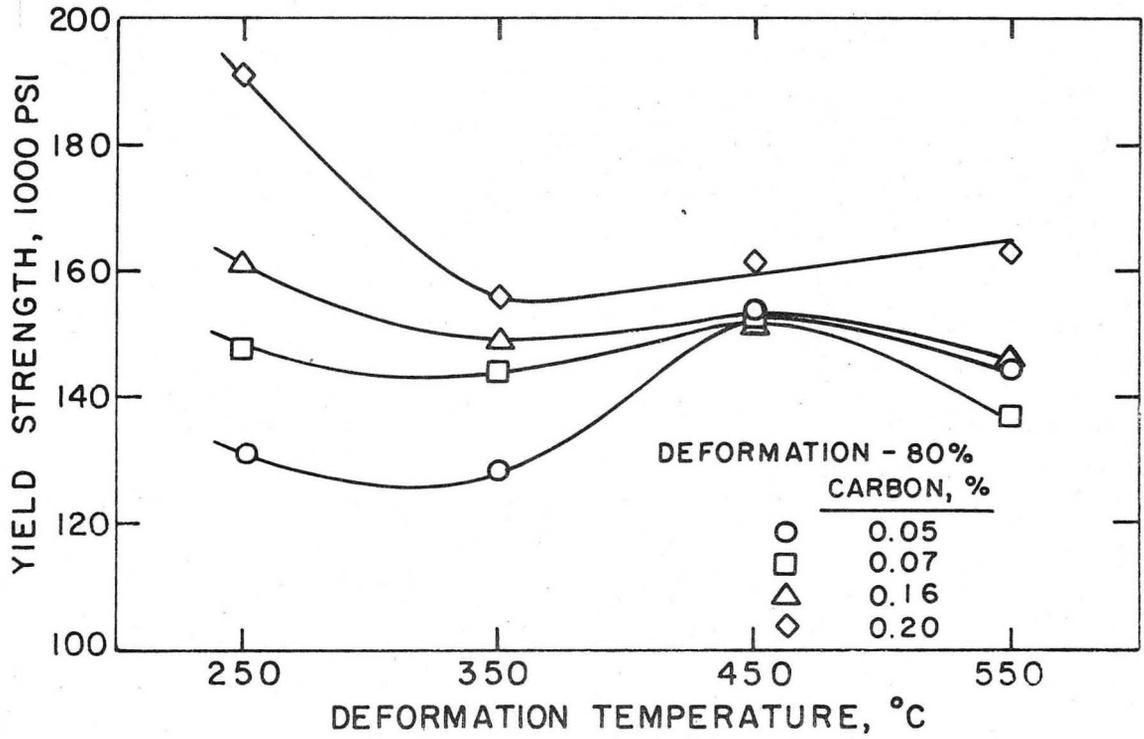
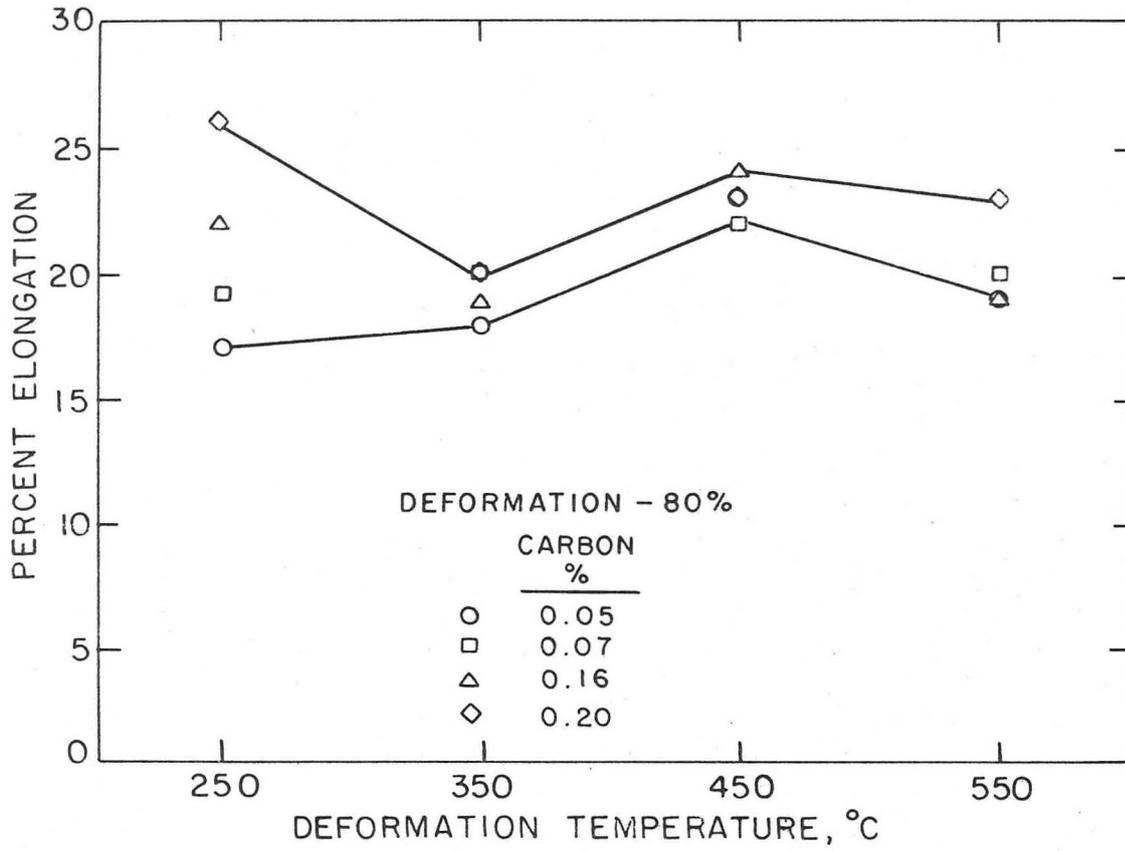


Fig. 3



XBL 697-1039

Fig. 4



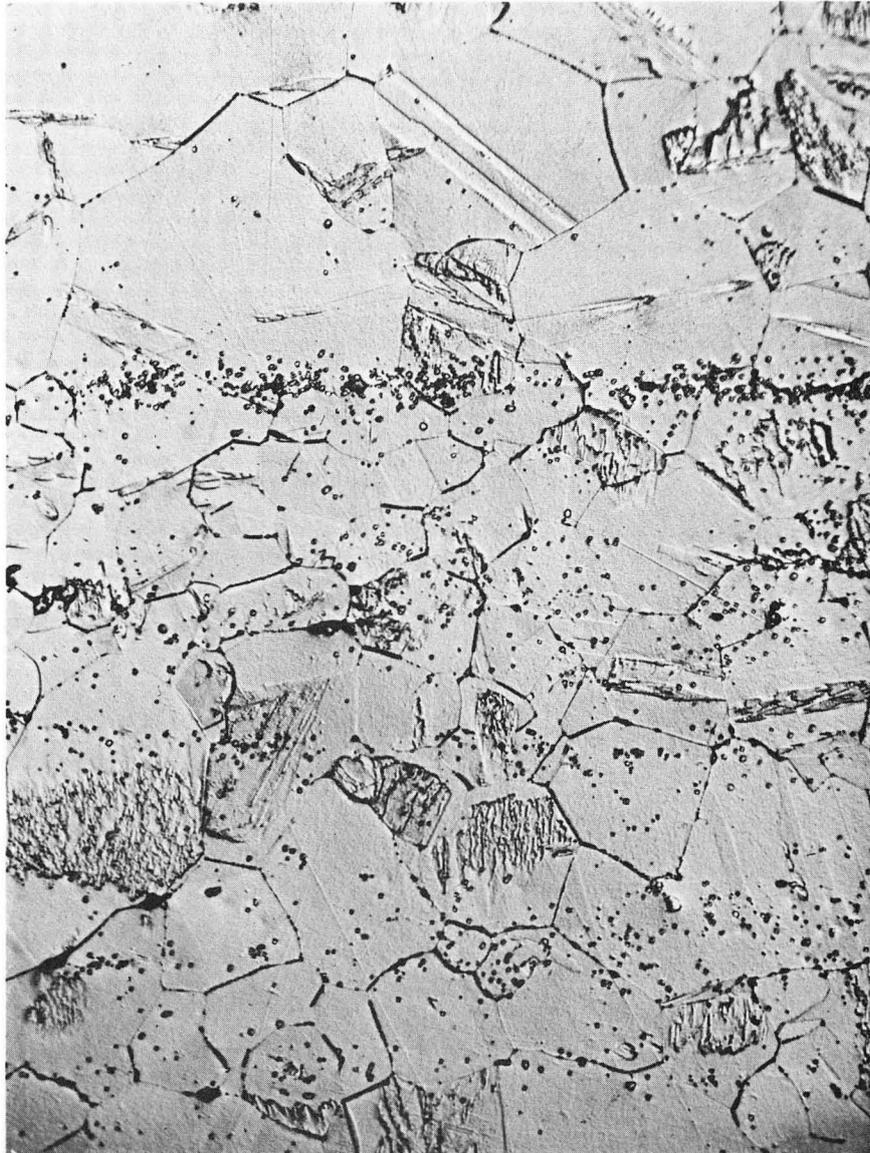
XBL 698-1196

Fig. 5



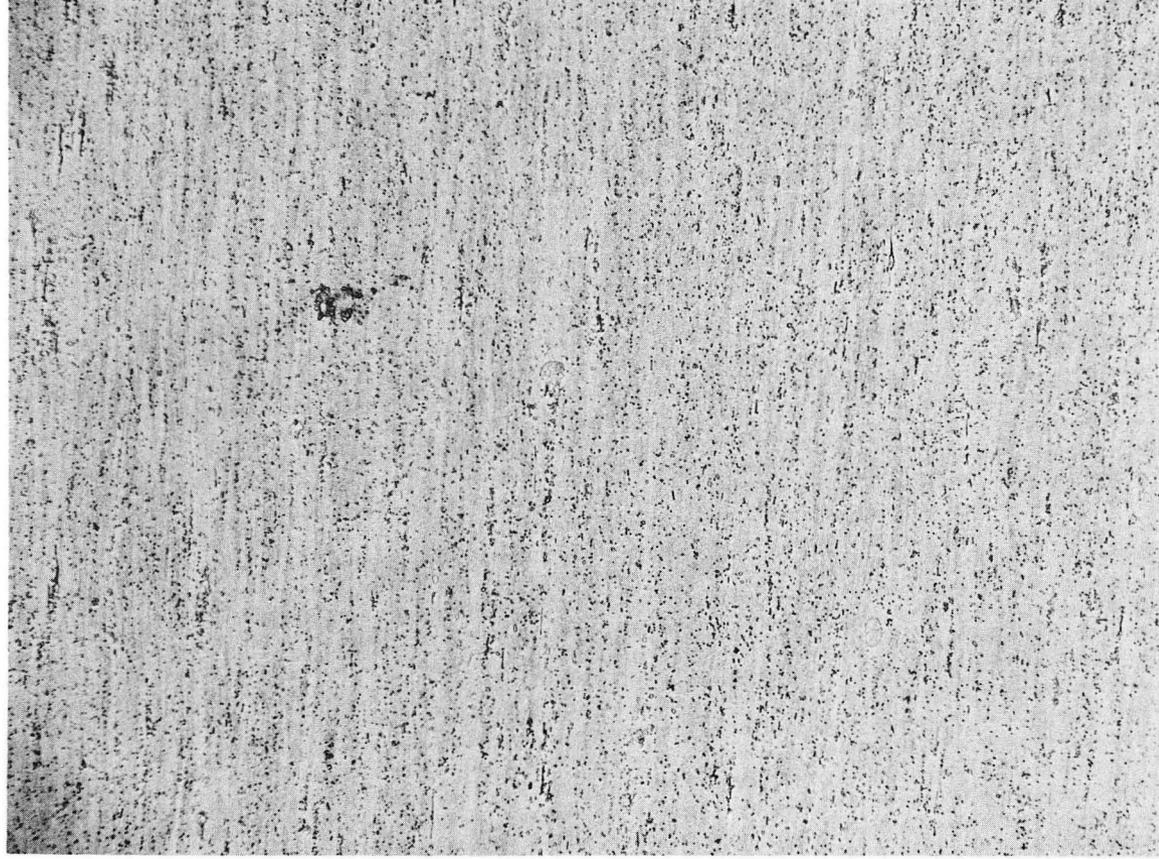
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Fig. 6



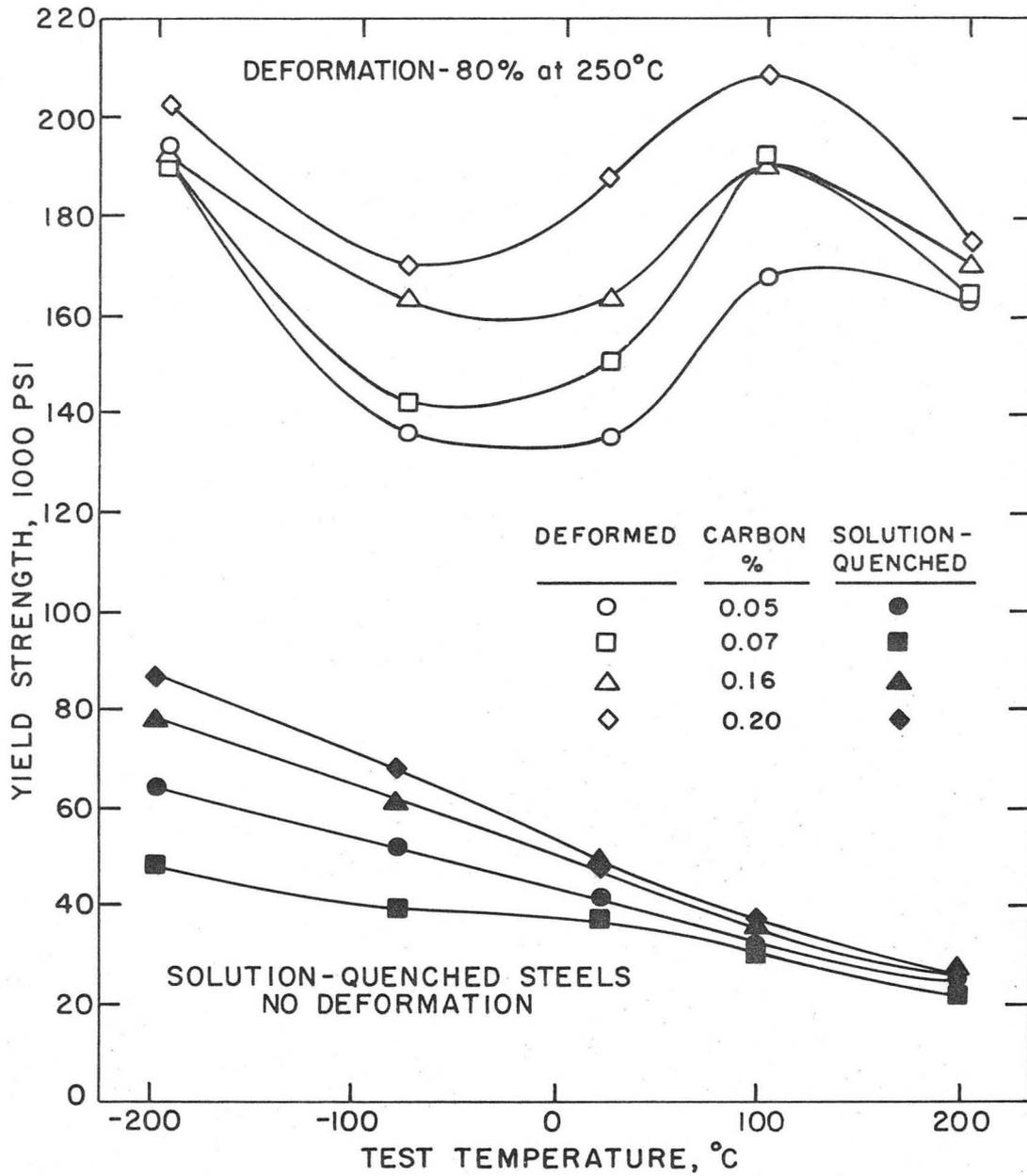
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Fig. 7



XBB 695-3026

Fig. 8



XBL 697-1040

Fig. 9

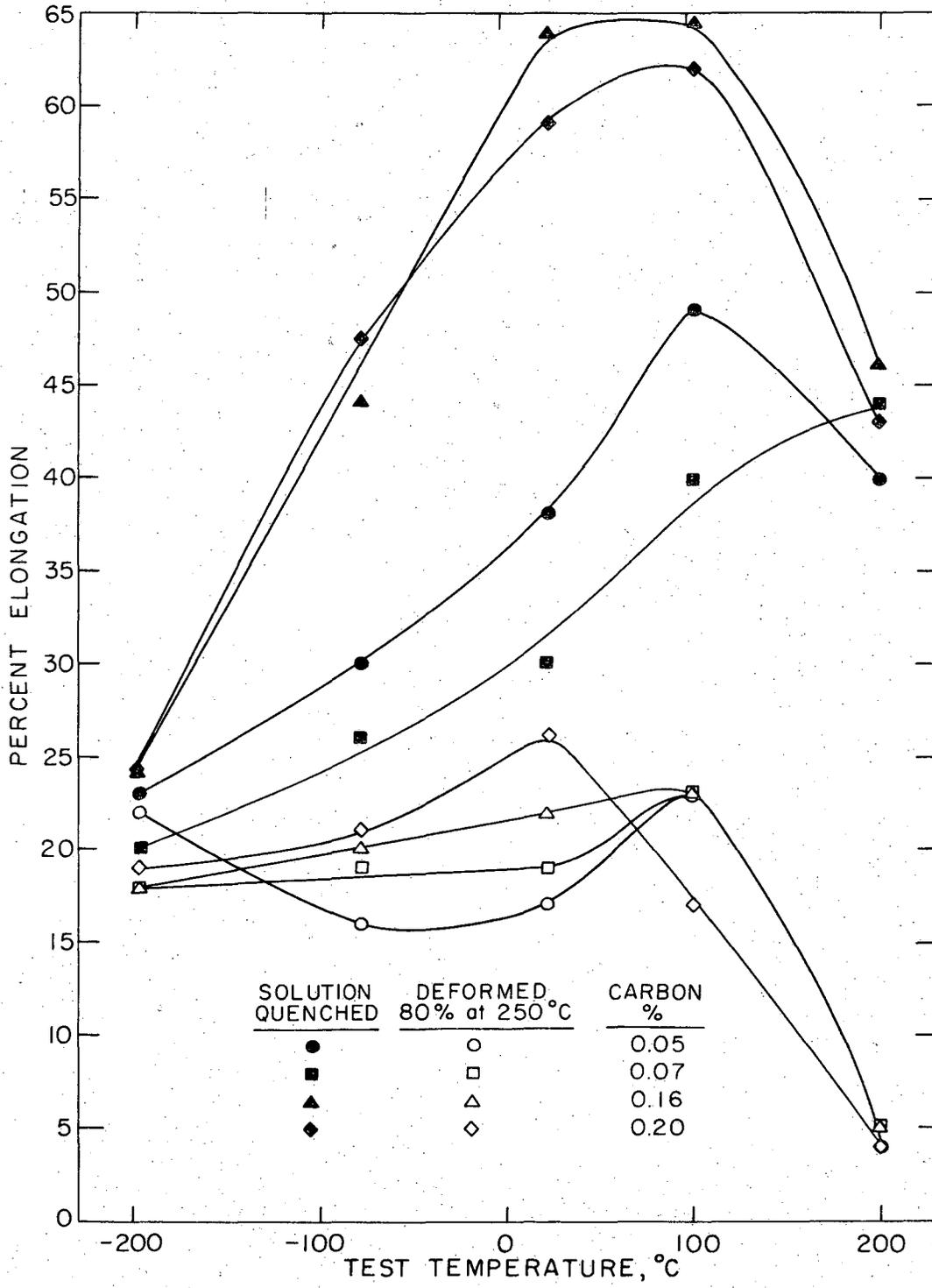
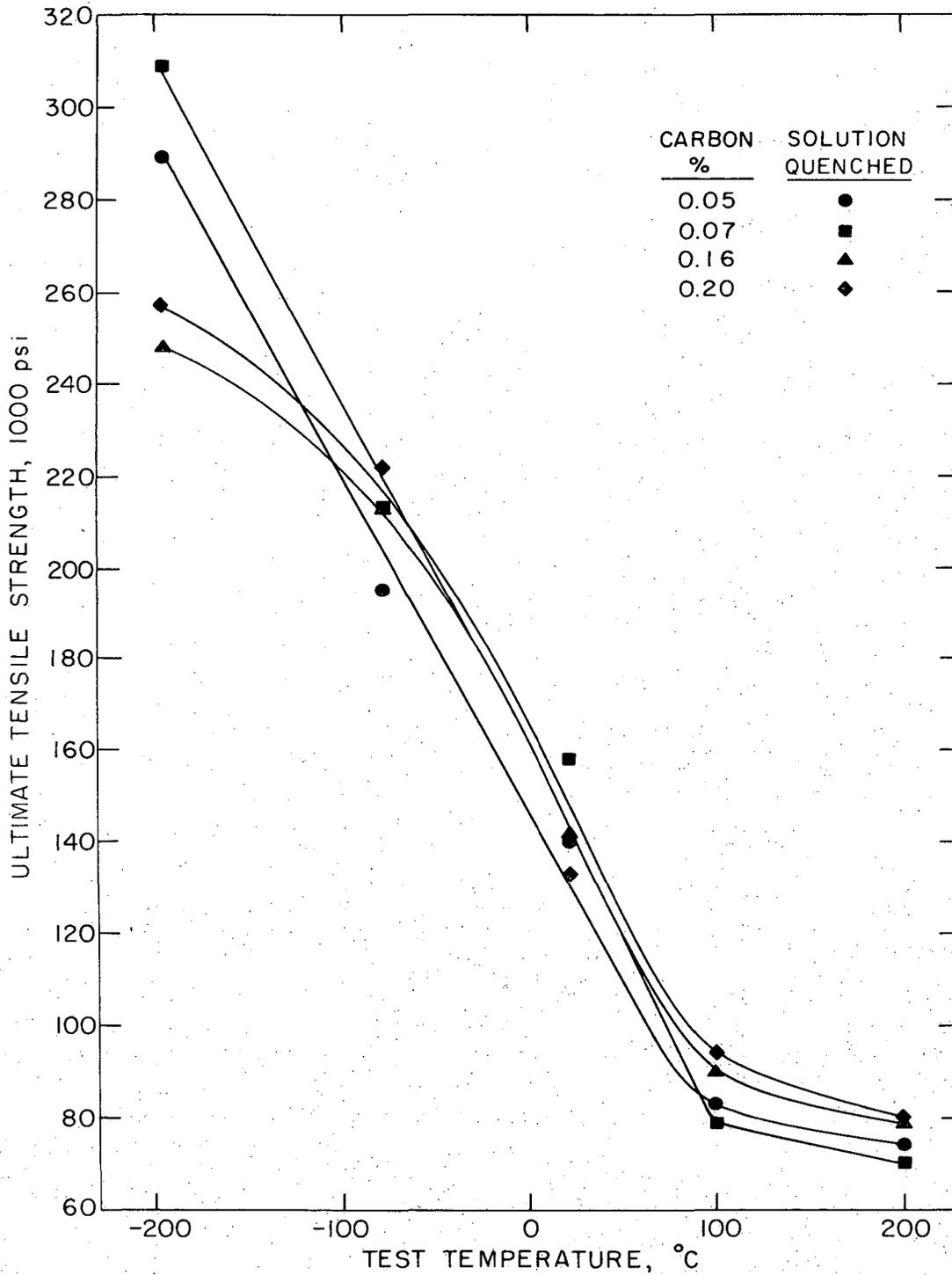


Fig. 10



XBL 689-5879

Fig. 11

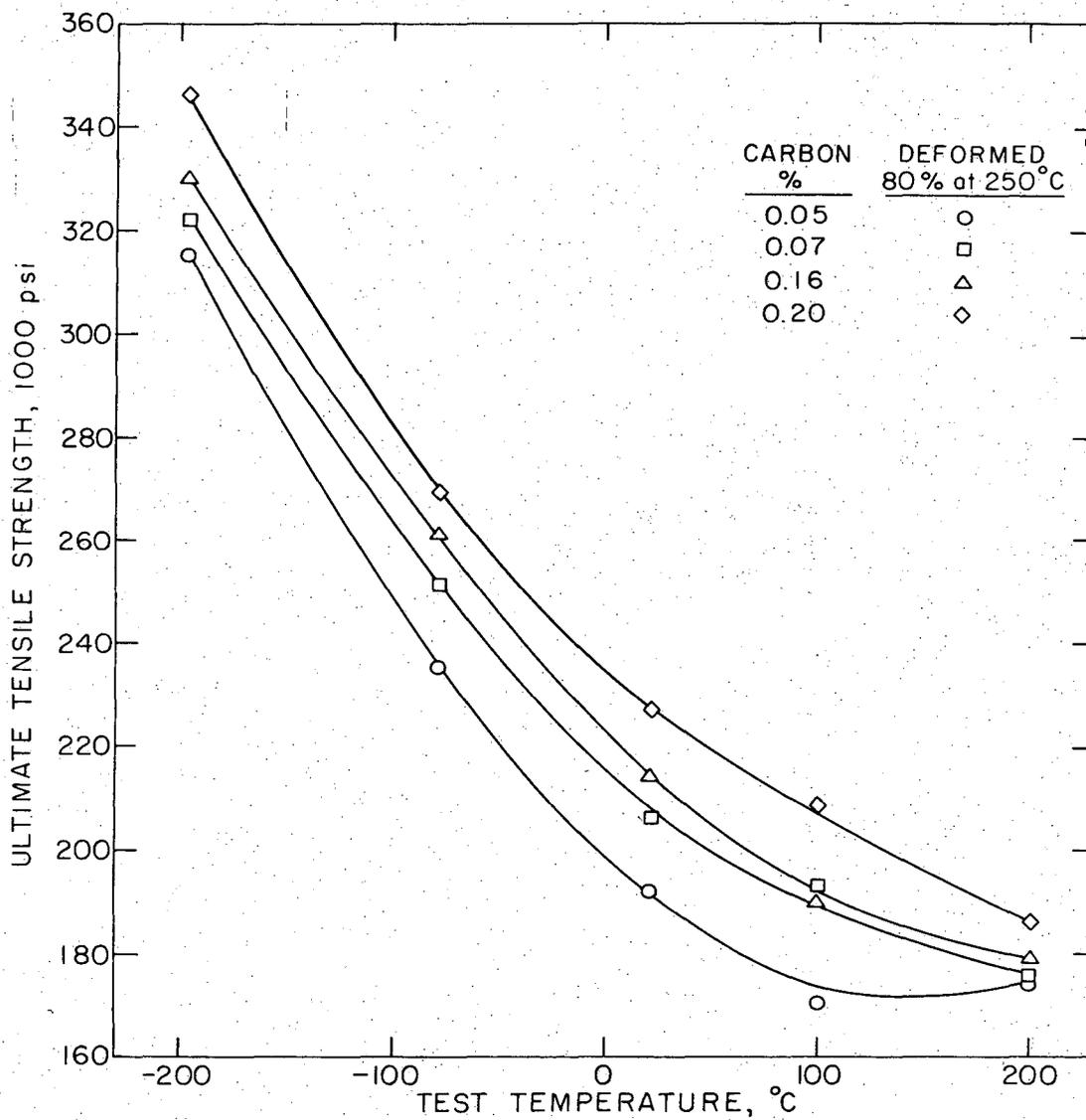
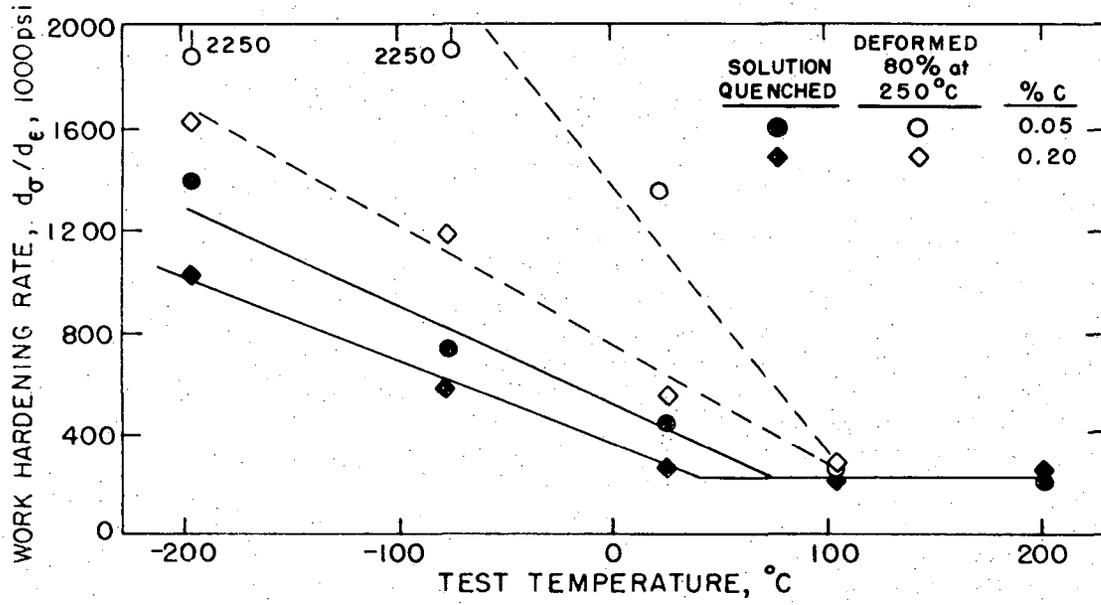
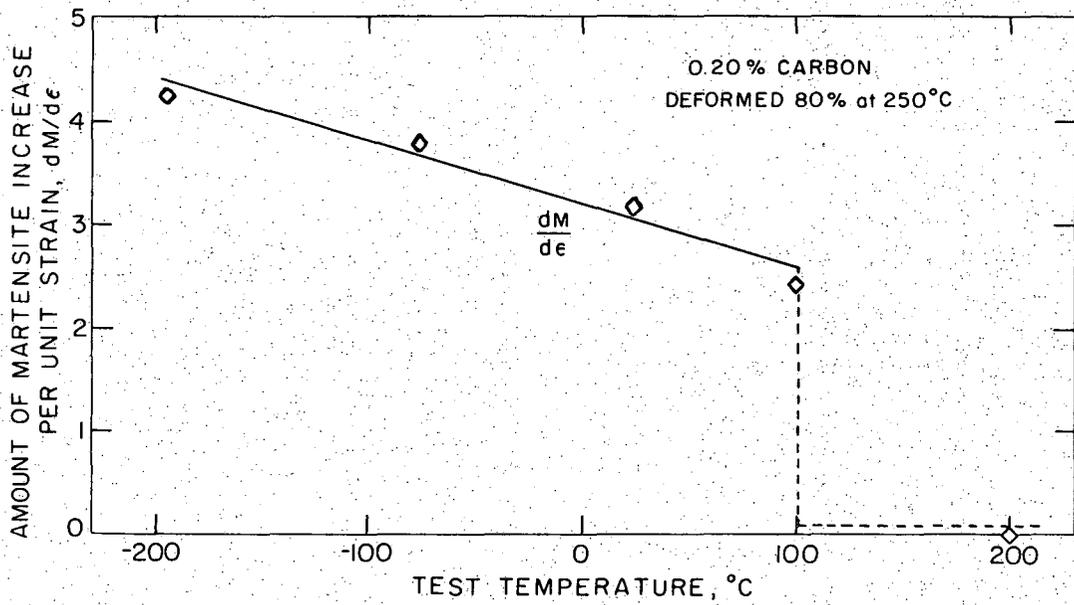


Fig. 12



XBL 695-499

Fig. 13



XBL 695-500

Fig. 14

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