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AN INVESTIGATION OF THE STRUCTURE AND PROPERTIES OF IRON-MANGANESE-CARBON ALLOYS

David George Atteridge
(D. Eng. thesis)

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AN INVESTIGATION OF THE STRUCTURE AND PROPERTIES OF IRON-MANGANESE-CARBON ALLOYS

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AN INVESTIGATION OF THE STRUCTURE AND PROPERTIES OF IRON-MANGANESE-CARBON ALLOYS

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ABSTRACT

The experimental research performed for this thesis resulted in development of an Fe-12 wt % Mn-0.2 wt % C high-strength TRIP steel with a yield strength of 180ksi, an ultimate strength of 240ksi, and an elongation of 20 percent. The processing consisted of a preliminary reduction at 750C, slightly into the γ single-phase region, followed by final warm rolling at 400C.

The feasibility of reducing both the alloy cost and the production cost of high-strength TRIP steel was demonstrated. Alloy cost can be reduced through the use of the Fe-Mn-C system, as the present TRIP steels contain both high-cost Ni and Mo. Production costs can be reduced by a decrease in the amount of warm-rolling reduction required to produce a given-yield-strength TRIP steel by the inclusion of a low-temperature hot-rolling step prior to warm rolling.

The research effort also included a set of experiments aimed at determining the feasibility of developing a low-cost, medium-strength TRIP steel (100-125ksi). It was found that one could achieve this by hot rolling a given Fe-Mn-C alloy at temperatures at the lower end of the γ single-phase region. A variety of alloy compositions were amenable to this type of processing and resulted in alloys with higher yield strengths, 100 vs. 60ksi -- and comparable elongations, 15 to 20 percent -- than the

present industrial hot-rolled alloys; the current hot-rolled alloys are all ferritic alloys, while alloys of the present investigation are austenitic as processed.

The results of this work established the feasibility of producing low-cost TRIP steels for both high and medium strength requirements. Thus the combination of the increased resistance to fracture demonstrated by TRIP steel and the improved resistance to hydrogen embrittlement and stress corrosion cracking of an austenitic matrix over that of a ferritic matrix is potentially within the grasp of industry.

I. INTRODUCTION

The industrial use of high-strength steel has been plagued with the problem of decreases in elongation and fracture toughness with increasing yield strength. Recently a new type of alloy, TRIP steel, was designed to eliminate both of the above-mentioned problems.[1] The acronym TRIP stands for Transformation Induced Plasticity, and it describes a class of metastable austenitic steels that exhibit a strain-induced phase transformation from γ to α'. The TRIP steels, which are austenitic until plastically deformed, exhibit yield strengths up to 250ksi, elongations from 20 to 40%, and fracture toughness values (in thin sheets) up to 500ksi-in.$^{\frac{1}{2}}$[1-11] And yet with all these assets TRIP steels remain, for the most part, only an interesting academic study in alloy design because problems related to their processing preclude their commercial use.

Two major stumbling blocks to increased industrial use of TRIP steels are the high cost of the alloying elements and the high cost of processing. A typical TRIP steel contains 9% Cr, 8% Ni, and 3.5% Mo plus small amounts of Mn, Si and C; Ni and Mo are particularly expensive. The processing requires that the alloy be given an 80% reduction in thickness by warm rolling at temperatures up to 450°C; this requires high-strength rolls and high-load-capacity mills. The objective of the present investigation was to apply alloy design theory to the problem of developing TRIP steels with substantially reduced cost, from the standpoint of both alloying elements and processing.

The basic requirement of a TRIP steel alloy system is that its alloy chemistry be suitably balanced to produce a metastable austenitic matrix.[3-5,12] This requires that at least one of the major alloying

elements be a strong austenite stabilizer: in the existing system both Ni and C are austenite stabilizers. A secondary requirement is that the system be capable of attaining a high austenitic yield strength. This is done in the existing system by combining carbide precipitation (due to the strong carbide formers Mo and Cr) with substructure refinement (through warm rolling).[13-21] It is felt that the warm rolling essentially eliminates any continuous high-angle grain boundary structure--leaving a very fine cellular substructure--, and that fine, coherent precipitation or clustering also takes place during the processing, resulting in high yield strengths. Warm rolling is used because one must be above the M_d temperature -- the temperature at which the matrix first becomes unstable with respect to plastic deformation -- in order to prevent the formation of α' during processing, and one must be sufficiently below the equilibrium γ range to avoid strain-induced recrystallization and, at the same time, at a low enough temperature to eliminate gross carbide precipitation.[22,23]

This type of steel demands that the phase stability of the given alloy system be balanced such that carbide precipitation take place in the γ matrix at warm-rolling temperatures and that the γ matrix be stable at room temperature with respect to athermal transformation to α', but unstable with respect to a strain-induced phase transformation. Yet the matrix must be sufficiently stable against this strain-induced transformation to resist transformation until plastic deformation has taken place; if the matrix is too unstable the transformation proceeds during elastic deformation of the γ matrix and the steel exhibits a premature, low-yield-strength phenomenon.[5]

The alloy chemistry and processing variables must also be controlled so that the resultant TRIP steel exhibits the plastic strain-induced phase transformation in the applicable temperature range.[3-5,24] The temperature of a TRIP steel in industrial use must be confined to a limited region between the M_s temperature, where the matrix starts to athermally transform to α', and M_d temperature, above which the matrix remains austenitic under plastic deformation. This limitation is due to the resultant mechanical properties being controlled by the amount of martensite formed per unit strain, and to the fact that this rate of transformation changes as a function of both temperature and strain rate, with the amount of transformation decreasing as either is increased. Therefore, the alloy chemistry or a given TRIP steel must be carefully balanced to assure that its austenite will have the correct amount of metastability, after processing, in its required industrial temperature range.

The resultant metastability of the austenite is a complicated function of alloying chemistry and processing variables. The usual TRIP steel processing, prior to warm rolling, that is now used is that of austenitizing the material at a high enough temperature of eliminate residual carbides, and then rapidly quenching to suppress the formation of equilibrium decomposition products. Both the optimum austenitizing temperature and the warm-rolling temperature, along with the optimum percent reduction in thickness, must be determined experimentally. The general mechanical property trends as a function of alloying content, and the processing variables for TRIP steels have been outlined,[3] but it is hard to relate these to equilibrium or metastable phase

diagrams because the phase relationships in multicomponent systems are either extremely complicated or unknown.

The work performed for this theses consisted of preliminary characterization of an alternative TRIP steel system, the Fe-Mn-C system, and development of methods for processing TRIP steels which are alternates to those now in use. The Fe-Mn-C system was chosen for several reasons. The as-solution-treated structure is known to change from α to γ with increasing Mn and/or C content, and carbides form in the γ matrix with increasing C content (Fig. 1).[25-27] Selected as-solution-treated alloys exhibit a γ-to-α' transformation upon straining.[28-30] Both C and Mn are low-cost alloying elements. In addition, several practical industrial alloys are based on this simple ternary system; austenitic manganese steel (Hadfield steel) is one example.

The Fe-Mn-C system is, therefore, a low-cost ternary alloy system which is known to exhibit a γ-to-α' phase transformation, with the possibility of the required γ-matrix carbide strengthening coming from Fe or Mn carbides. This system also held out the promise of the resultant mechanical properties being directly relatable to known equilibrium or metastable phases present at the given processing temperatures, as a relatively large amount of accumulated phase-diagram knowledge is available for this system.[25,29,31,32] The processing of this alloy system was guided by the known phase stability of the Fe-Mn-C system, and the desire to relate the resultant properties to it.

The experimental results of this investigation are categorized by processing procedure. The first section covers the mechanical properties

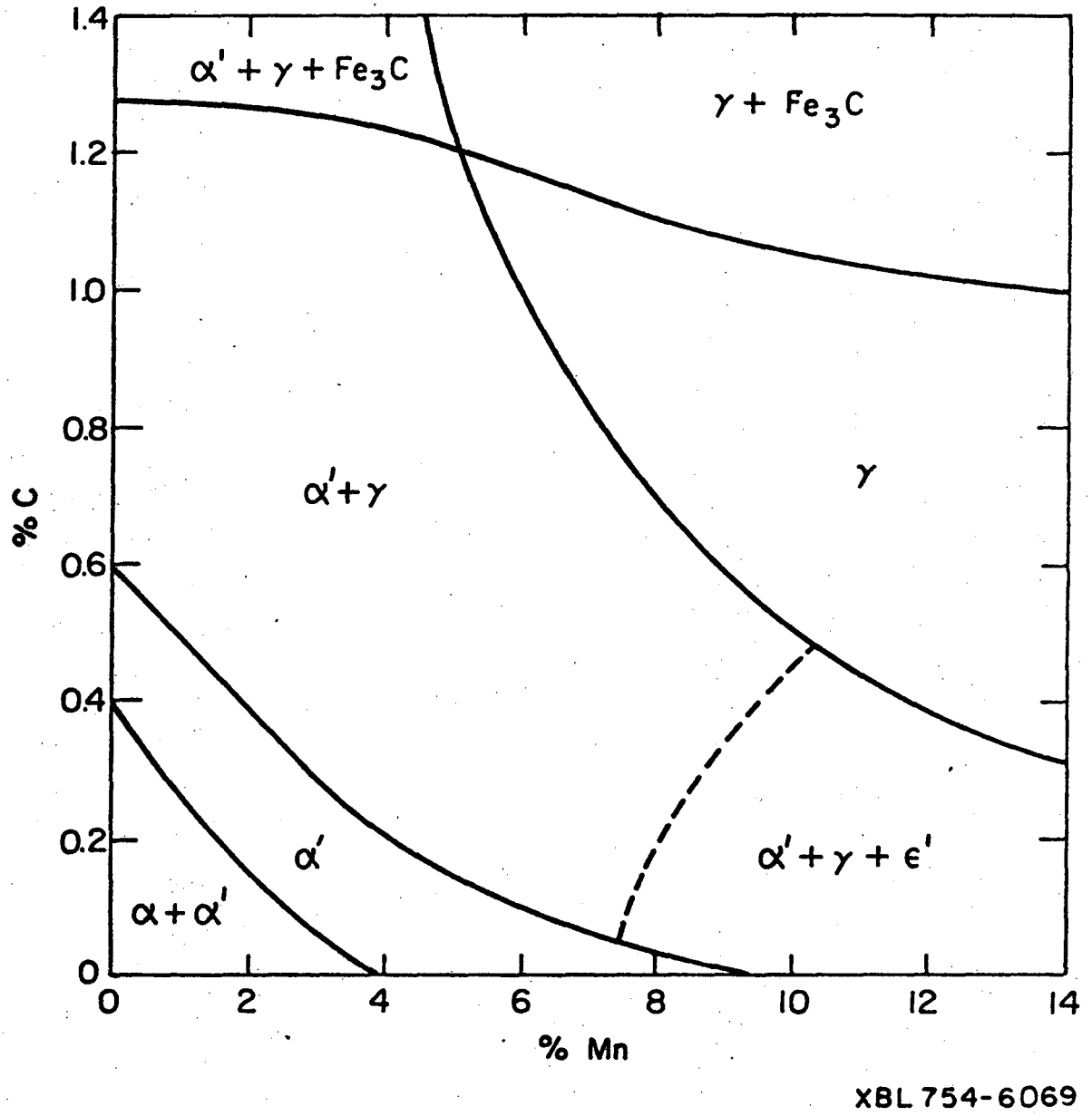

Fig. 1. Phases present in Fe-Mn-C system as solution treated.

of alloys given a modified hot-rolling treatment. The second section contains the mechanical properties of the warm-rolled alloys and the results of an investigation into modified warm-rolling procedures.

II. EXPERIMENTAL PROCEDURES

A. Alloy Processing

The raw materials used in preparing the alloys for this program were 99.99% pure. The selected materials were induction melted under an argon atmosphere and cast into water-cooled copper molds. The ingots were approximately 1" in diameter and weighed 1.5 lbs.; the alloy compositions and alloy designations used in this study are summarized in Table I.

The alloy ingots were doublebagged in stainless steel foil and homogenized at 1200C for 2 hours and then ice-brine quenched. This treatment was considered sufficient as there are no high-temperature carbides formed in the Fe-Mn-C system, and the recommended solution heat-treatment for as-cast Hadfield steel, which has a higher carbon content than the alloys used in the present study, was 1010C for 1/2 hour followed by a rapid quench.[25,33]

The alloys were cut into 1/4" slices and processed by either hot or warm rolling, or both, to yield a reduction of thickness of approximately 80%, unless otherwise indicated. The specimens to be hot rolled were again doublebagged in stainless steel foil and austenitized at 1100C for 1/2 hour; they were then removed from the furnace and placed directly into another furnace at the required hot-rolling temperature. The bags were removed from the specimens just before they were placed in the hot-rolling furnace; the hot-rolling temperatures varied from 650C to 950C.

TABLE I. Alloy compositions and designations.

Alloy Designation	Alloy Content, wt %	
	Mn	C
12-0	12	0
12-3	12	0.20
12-5	12	0.52
12-8	12	0.82
12-10	12	1.09
10-3	9.8	0.35
10-5	9.8	0.48
10-8	9.9	0.79
10-10	9.9	1.07
8-3	8.0	0.35
8-5	7.8	0.50
8-8	8.0	0.81
8-10	8.1	1.03
6-5	6.2	0.47
6-8	6.3	0.85
6-10	6.3	1.10

through a multipass rolling schedule. The alloys were replaced in the furnace between passes to assure a constant rolling temperature. The rolls were preheated to 150C to minimize cold-roll surface-quenching effects; this was done for the warm-rolled specimens also. The specimens were water quenched after the final reduction pass.

The warm-rolled specimens were given a prior hot-rolling treatment essentially the same as that outlined above except that the percent thickness reductions were varied, as mentioned in the Experimental Results section. The warm-rolled specimens were heated to the warm-rolling temperature from room temperature, and were warm rolled down through a multipass process to yield a total reduction of 80% calculated from the pre-hot-rolling thickness. The warm-rolled specimens were replaced in the furnace between passes, as in the hot rolling; warm-rolling temperatures varied from 200C to 500C. Processing deviations from the general procedures outlined above are mentioned in the text.

B. Mechanical and Magnetic Testing

Tensile specimens were machined from the as-rolled material; the specimens were approximately 0.05 in. thick and were machined to the configuration given in Fig. 2. The finished specimens were tested in an Instron testing machine using a crosshead speed of 0.1 cm/min.

The volume percent martensite in the specimens was determined through the measurement of the specimen's saturation induction by a permeameter; the permeameter was designed for use with this specimen configuration, and its design and use are discussed elsewhere.[34] The effect of the various alloying elements on the saturation induction of iron was determined from the literature, and was used to calculate the

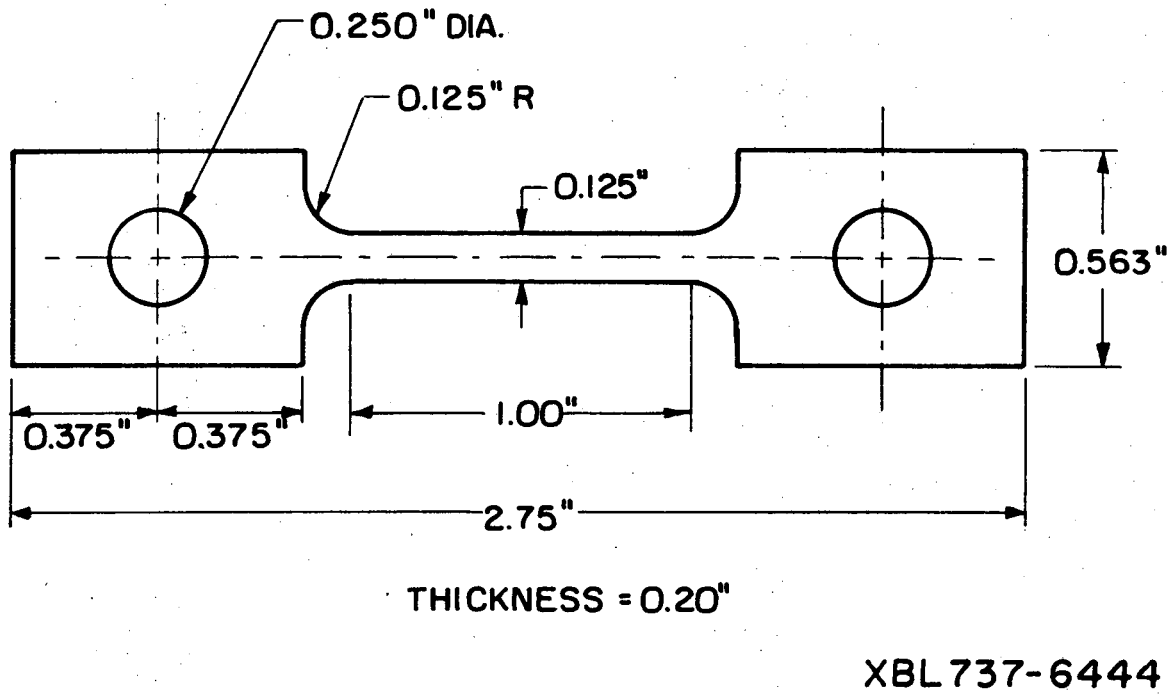

Fig. 2. Dimensions of flat tensile test specimen.

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percent α' present in the given sample prior to and following mechanical testing.[25,35,36] The mechanical property results and pre-and post-test percent of magnetic phase present in the specimens, and the average α' formation rate (in volume percent α' per unit strain) are summarized in Tables II through VIII.

TABLE II. Mechanical roperties for alloys rolled at 250C, 450C and 650C.

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
		250C				
12-0	186	189	3	98	99	0.33
		450C				
12-0	159	163	3	94	99	1.67
		650C				
12-0	78	133	10	67	88	2.10
	124	131	7			
12-3	60	174	21	0	62	2.95
	69	173	19			
12-5	108	142	11	0	4	0.36
	109	152	13			
12-8	133	160	6	0	4	0.67
	138	178	10			
12-10	159	188	4	0	5	1.25
	156	202	7			

TABLE III. Mechanical properties for alloys rolled at 750C.

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
12-0	79	125	8	66	89	2.88
	66	127	10			
12-3	46	171	19	0	65	3.42
	57	157	18			
12-5	101	148	14	0	5	0.36
	96.5	138	12			
12-8	128	162	8	0	5	0.63
12-10	144	180	6	0	4	0.66
	138	172	6			
10-3	74	89	3	2	24	7.35
	75	96	6			
10-5	81	111	10	0	8	0.80
	87	176	12			
8-3	121	>185	>1	30	73	43
	93	>224	>7			
8-5	74	96	3	0	24	8
	78	101	5			

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TABLE IV. Mechanical properties for alloys rolled at 850C.

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
12-3	71	224	21	0	92	4.38
	69	208	17			
12-5	104	158	22	0	10	0.46
	109	151	17			
12-8	127	167	7	0	5	0.71
	129	173	11			
12-10	140	181	6	0	4	0.66
	147	187	5			
10-3	85	177	13	1	51	3.80
	88	152	11			
10-5	113	166	13	2	28	2.0
	120	154	11			
10-8	110	150	5	0	3	0.60
	102	147	6			
10-10	135	176	5	0	4	0.80
	141	177	4			

TABLE IV. Mechanical properties for alloys rolled at 850C (Cont.)

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
8-3	111	220	>4	38	91	13.25
	*	223				
8-5	78	166	7	6	44	5.44
	74	150	7			
8-8	121	144	5	5	7	0.40
	115	142	6			
8-10	128	171	3	5	7	0.66
	134	168	4			
6-5	152	264	5	53	100	9.40
6-8	99	112	4	0	8	2.0
	91	119	5			
6-10	127	158	3	2	17	5.0
	138	170	4			

*Not recorded.

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TABLE V. Mechanical properties for alloys rolled at 950C.

Alloy	Strength, ksi		%	% Martensite		
	Yield	Ultimate	Elongation	Pre-test	Post-test	% Formed/ % Elongation
12-3	46	190	>18	5	99	5.23
	50	221	23			
12-5	76	136	19	3	10	0.90
	101	149	16			
12-8	115	166	19	0	3	0.16
	118	170	19			
12-10	144	190	12	0	4	0.33
	139	187	14			
10-3	84	133	8	2	39	4.53
	78	176	11			
10-5	104	153	16	4	34	1.88
	111	152	15			
10-8	101	136	9	0	4	0.45
	115	148	8			
10-10	126	160	10	0	4	0.40
	116	151	11			

TABLE V. Mechanical properties for alloys rolled at 950C (Cont.)

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
8-5	85	175	6	8	43	5.84
	66	147	6			
8-8	102	132	8	5	8	0.38
	93	121	8			
8-10	118	142	4	1	4	0.75
	121	143	2			
6-5	153	265	5	56	100	8.80
6-8	93	141	5	10	20	0.50
	93	117	4			
6-10	131	149	5	0	7	1.40
	117	141	3			

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TABLE VI. Mechanical properties for alloys with preliminary reduction at 950C followed by secondary reduction at 400C.

Alloy	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
12-3	113	240	20	2	86	4.20
	113	235	23			
12-5	161	190	9	0	4	0.23
	164	197	8			
12-9	187	203	5	0	3	0.60
	216	236	9			

TABLE VII. Mechanical properties for alloy 12-3 given a preliminary reduction of 62 percent at 750C.

Secondary Rolling Temperature	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
500	133	220	24	0	79	3.29
	138	250	23			
400	138	223	18	0	92	5.11
	140	246	21			
400*	160	237	18	0	65	3.61
	156	222	16			
300	151	231	16	0	68	4.25
	155	228	20			
200	128	223	20			

*Continuous rolling procedures. See text.

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TABLE VIII. Mechanical properties for alloy 12-3 rolled 94 percent with preliminary reduction at 750C followed by secondary reduction at 400C.

% Reduction at 750C	Strength, ksi		% Elongation	% Martensite		
	Yield	Ultimate		Pre-test	Post-test	% Formed/ % Elongation
76	174	232	16	0	64	4.0
	178	238	15			
85	162	216	15	0	61	4.06
	160	220	14			
93	143	221	18	0	65	3.52
	136	216	17			

III. EXPERIMENTAL RESULTS

This section breaks down the experimental results into two sub-sections. Each sub-section reports on a different processing procedure. The first contains the results of the hot-rolled alloys; the second contains the results of the warm-rolled alloys.

A. Hot-Rolled Alloys

The alloy compositions for these steels were selected so as to straddle the as-solution-treated $\alpha' + \gamma$ to γ boundary, as defined in Fig. 1. The alloys ranged in composition from 6 to 12 wt% Mn and from 0 to 1.1 wt% C (Table I).

The thermomechanical processing for these alloys began with a high-temperature austenitizing treatment at 1200C followed by a rapid quench. Then the alloys were given a medium-temperature austenitizing treatment at 1100C and put directly into the hot-rolling furnace. The selection of the hot-rolling temperatures for each alloy was based upon the equilibrium phase diagram for the given alloy. The diagrams used as a basis for this selection are given in Figs. 3-6, where the phase boundaries are plotted as a function of changing C content for a given Mn content.[25]

The experimental results for the hot-rolled alloys are broken down first by Mn content and then by carbon content; a composite overview is given in the last section on hot-rolled alloys.

1. Fe-12% Mn Alloys

The as-solution-treated structures for these alloys were expected to vary from a mixture of α', ε' and γ for the 12-0 and 12-3 alloys through the completely γ structures of the 12-5 and 12-8 alloys to the γ+C structure of the 12-10 alloy, as predicted in Fig. 1. The presence of the a' phase in the first two alloys was confirmed through the use of a hand magnet;

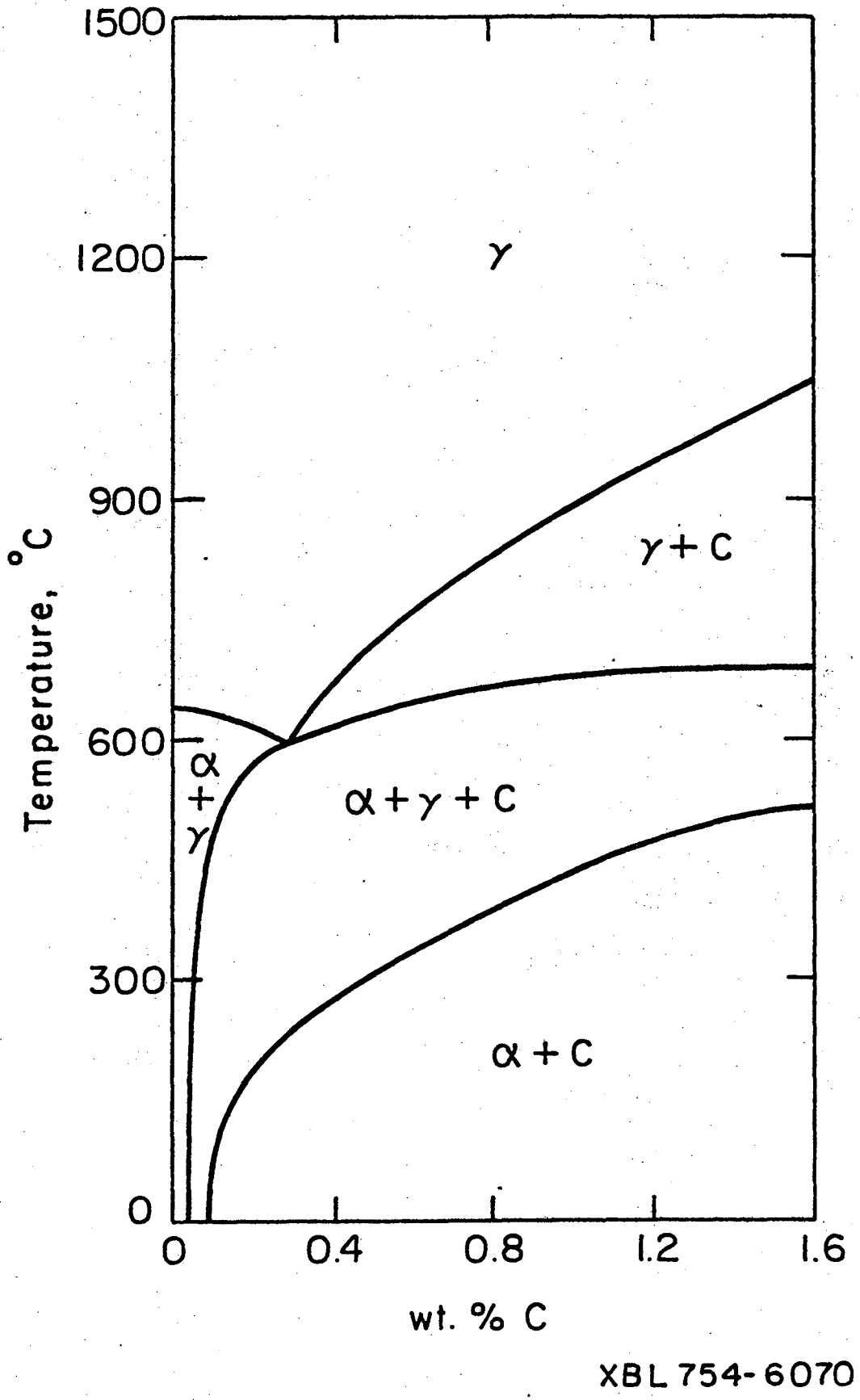

Fig. 3. Equilibrium phase diagram for Fe-12%Mn-X%C.

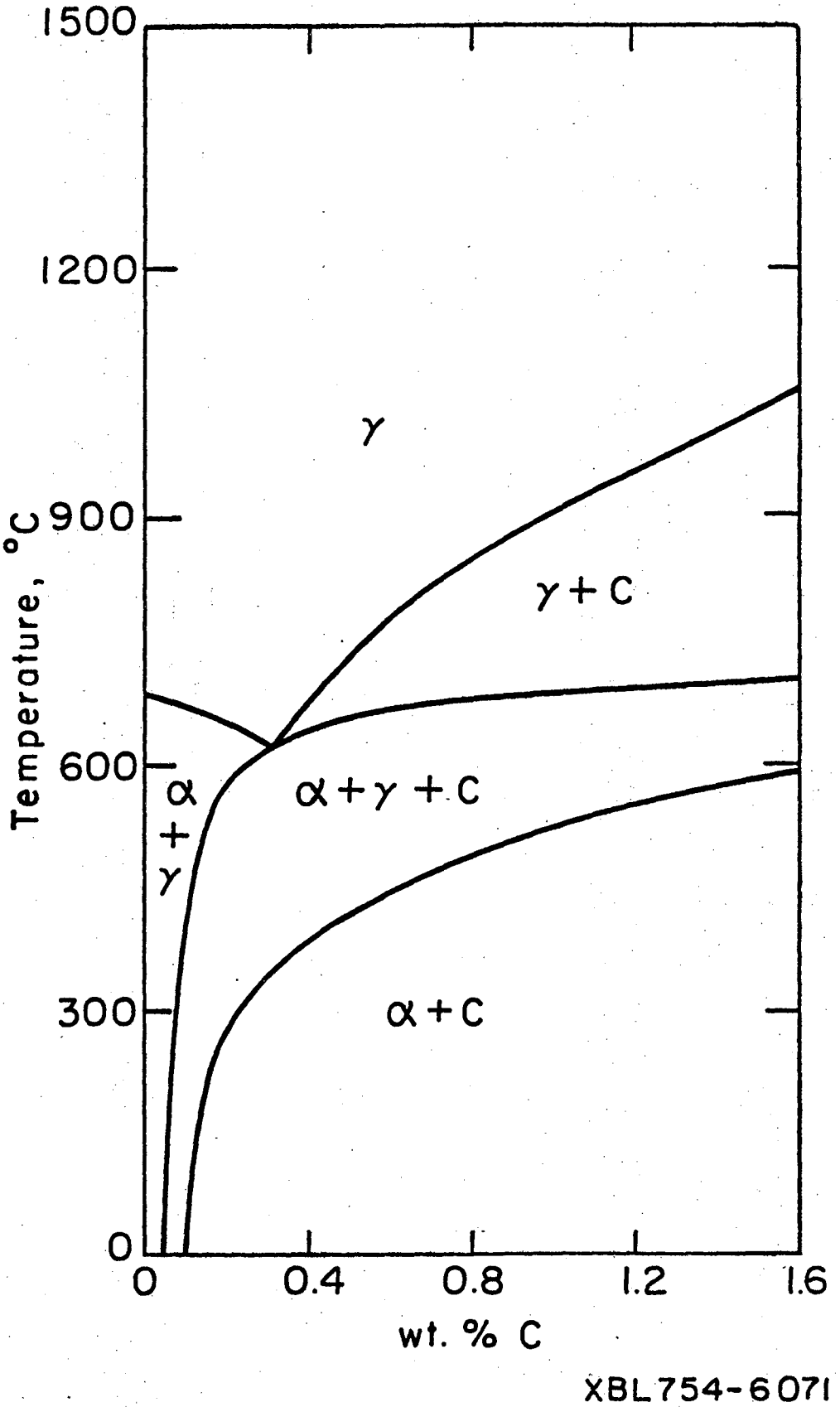

Fig. 4. Equilibrium phase diagram for Fe-10%Mn-X%C

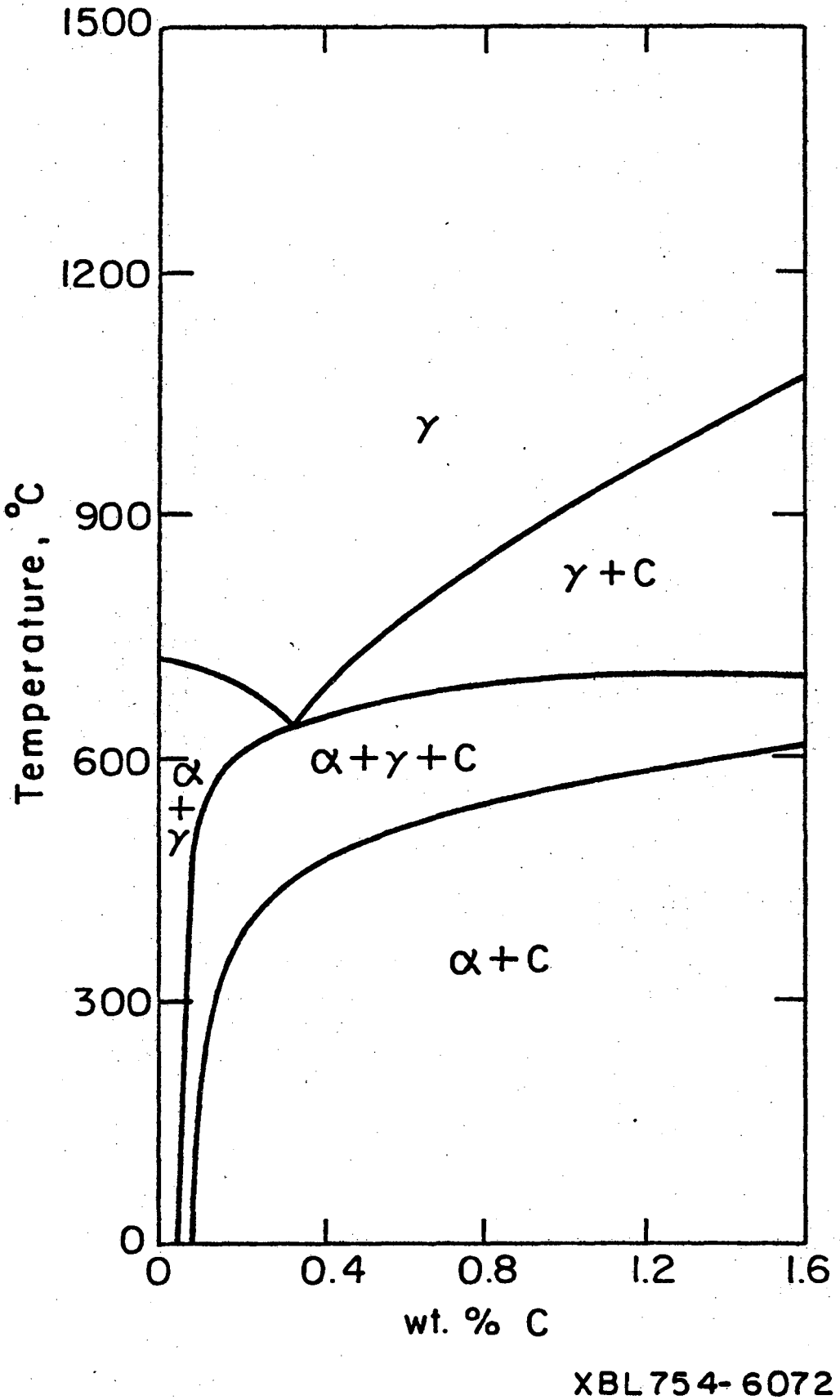

Fig. 5. Equilibrium phase diagram for Fe-8%Mn-X%C.

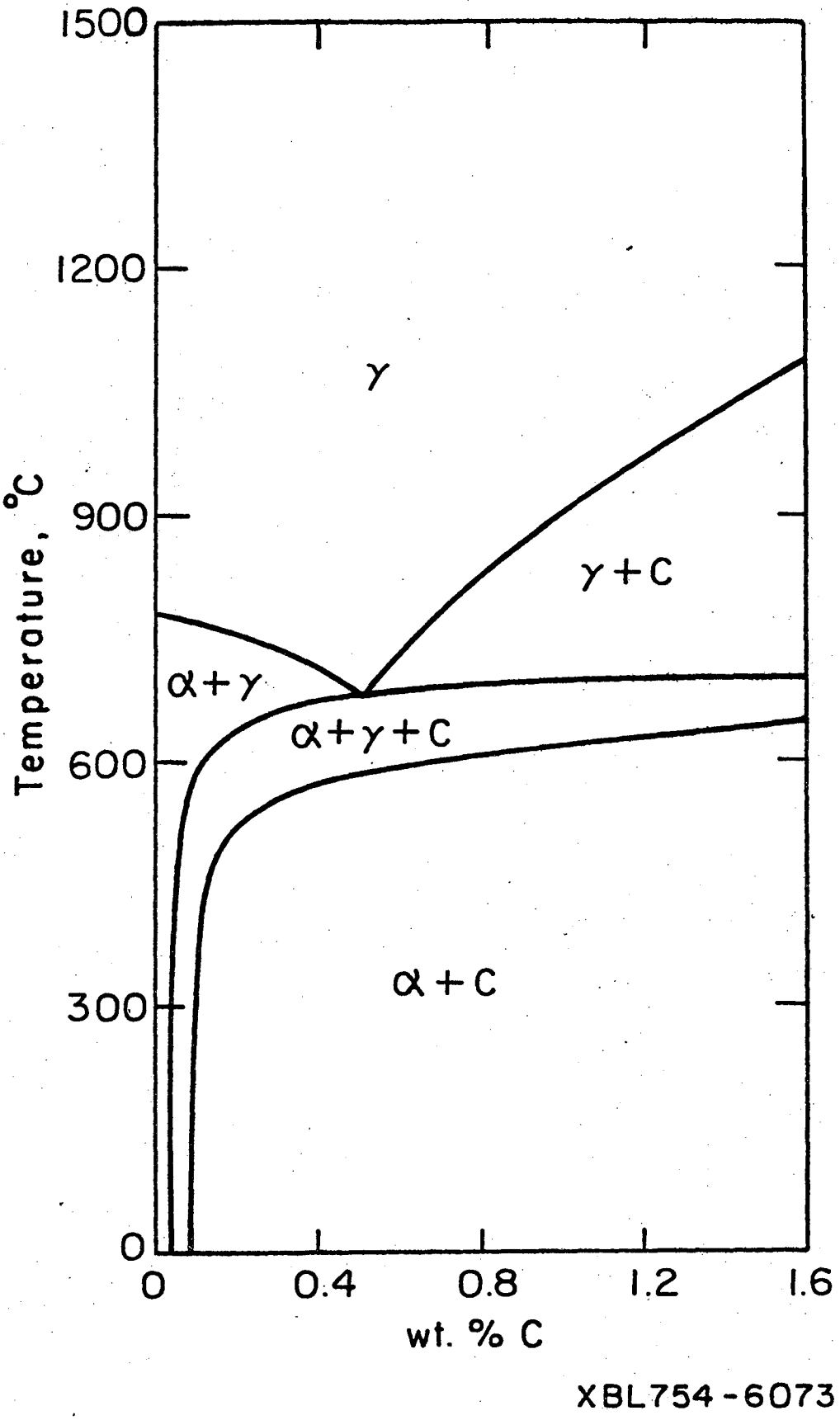

Fig. 6. Equilibrium phase diagram for Fe-6%Mn-X%C.

the other alloys were found to be essentially nonmagnetic using this same test. This series of alloys was then rolled in various regions of the equilibrium phase diagram, as determined in Fig. 3, in an attempt to determine the effect of the change in rolling temperature (and hence in the resultant mechanical and magnetic properties) in terms of the equilibrium phase stability at the given rolling temperatures.

The 12-0 alloy was prepared to allow determination of the effect of the change in amount of α' formation as a function of rolling temperature without the complication of having carbide precipitation change the alloy content (and therefore the phase stability) of the matrix. This alloy was rolled in both the γ and the $\gamma+\alpha$ regions of the equilibrium phase diagram; the resultant mechanical properties are summarized in Fig. 7 with the equilibrium structure at the respective rolling temperature given at the top of the figure. It can be seen that the yield strength decreases as the rolling temperature increases, and decreases dramatically once the rolling is done in the γ range. Both the amount of strain hardening--as indicated by the difference between the yield and ultimate strengths--and the elongation increase as the rolling temperature increases.

These property changes correspond directly to the change in the percent of α' present in the alloy prior to and during testing. The percent of α' present prior to testing is proportional to the yield strength and decreases as the rolling temperature increases; it decreases substantially, as does the yield strength, once the rolling is done in the γ region. The rate of formation, and the total amount of α' formed during the tests, also changes as a function of rolling temperature, with the average amount of α' formed per unit strain increasing as the rolling temperature increases, while the total amount found at fracture decreases (Fig. 8). The average rate of α'

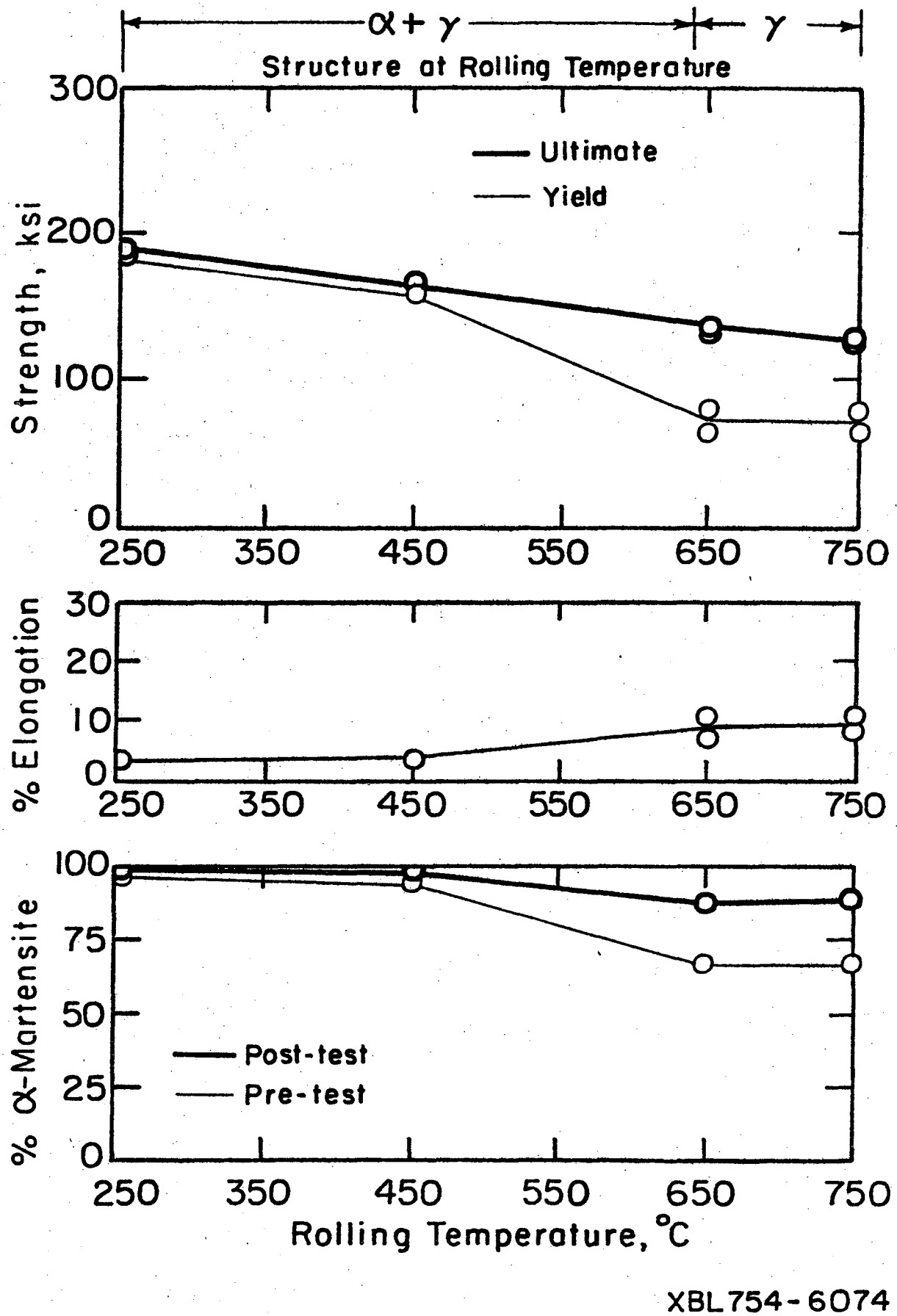

Fig. 7. Mechanical properties for 12-0 alloy.

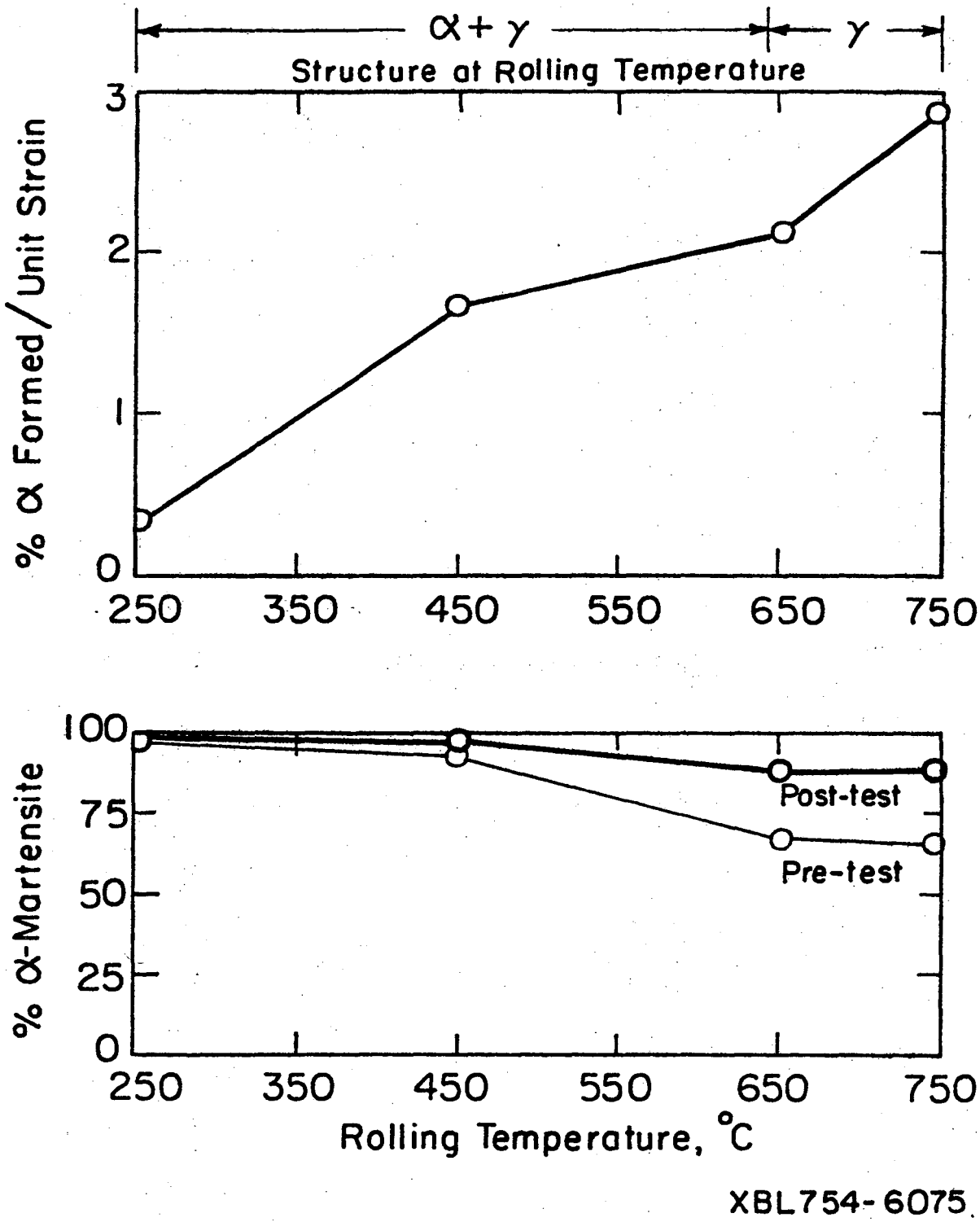

Fig. 8. γ to α transformation characteristics for 12-0 alloy.

formation is essentially a qualitative measure of γ resistance to strain-induced transformation, as the actual transformation rate varies as a function of plastic strain.[3,37] Nevertheless, it is felt that this data analysis technique is a good comparative indicator of relative γ stability.

The reduction of the amount of α' present after rolling in the γ region was found to carry over to the other carbon alloys; this was strikingly evident in the 12-3 alloy. This alloy was magnetic, as determined using a hand magnet, in the as-solution-treated condition, but after rolling in the γ region was found to be nonmagnetic by the same test. Thus, hot rolling this alloy increased the stability of the γ matrix with respect to athermal transformation to α'. This increased stability was enough to yield a metastable γ alloy at room temperature. The stability of γ through plastic deformation has also been documented elsewhere.[38,39]

The resultant metastable γ alloy was found to exhibit a high degree of strain-induced transformation to α' during testing, as seen in Fig. 9. This resulted in high elongations and a high degree of work hardening. The yield strengths, elongations, and percentage of α' prior to testing remain approximately constant for all rolling temperatures, although it appears that the yield strength may be generally decreasing slightly with increasing rolling temperature. The matrix as-rolled is essentially retained γ, but both the work-hardening rate and the total percentage of α' after testing increase as the rolling temperature increases. This increase in the percentage of α' formed during testing and the increase in the average rate of formation (Fig. 10) is attributed to a decrease in the disorder of the retained γ matrix as the rolling temperature is increased.

The increase in austenite stability -- with respect to strain-induced transformation -- found with increasing matrix disorder (lower rolling

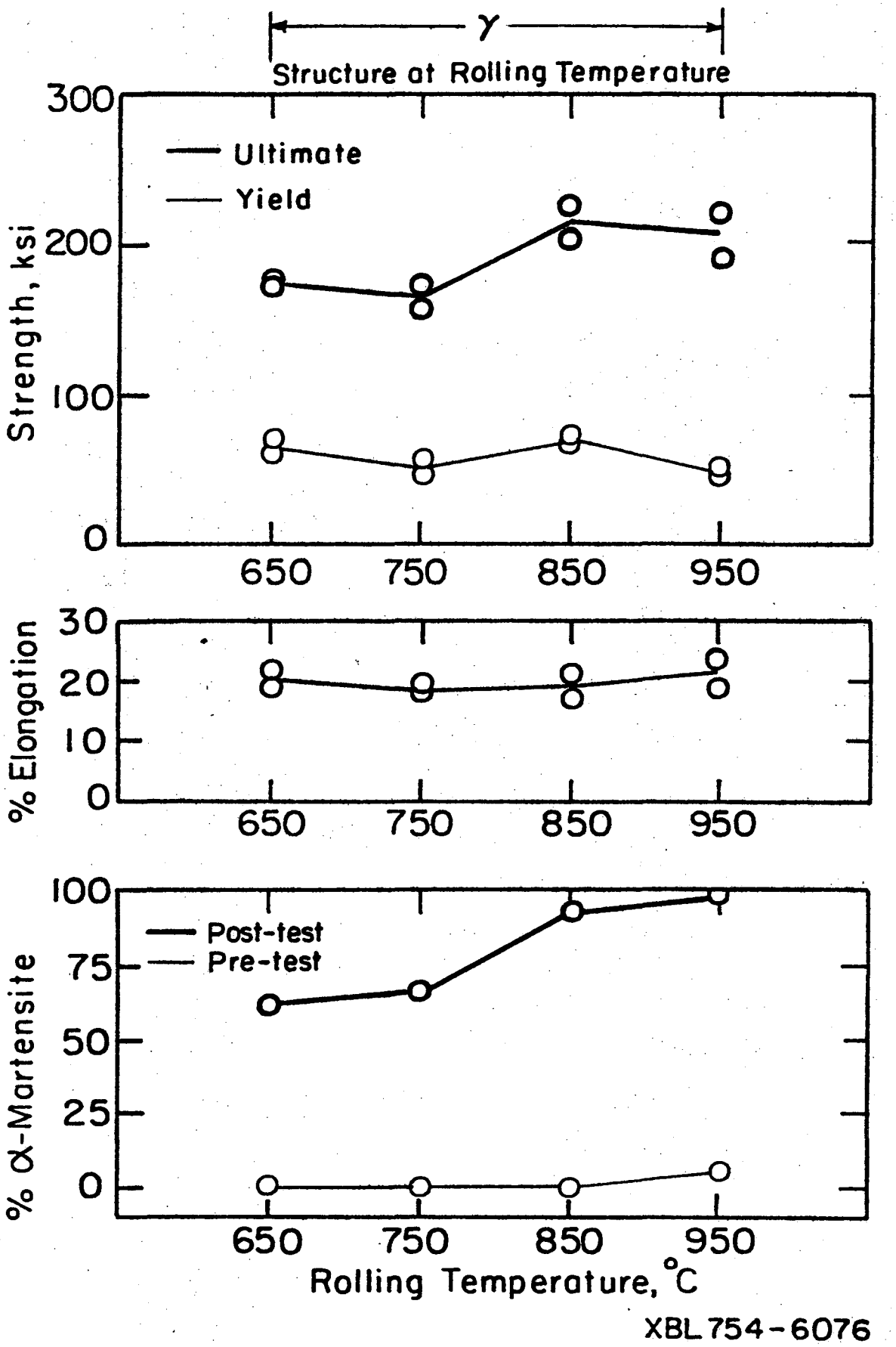

Fig. 9. Mechanical properties for 12-3 alloy.

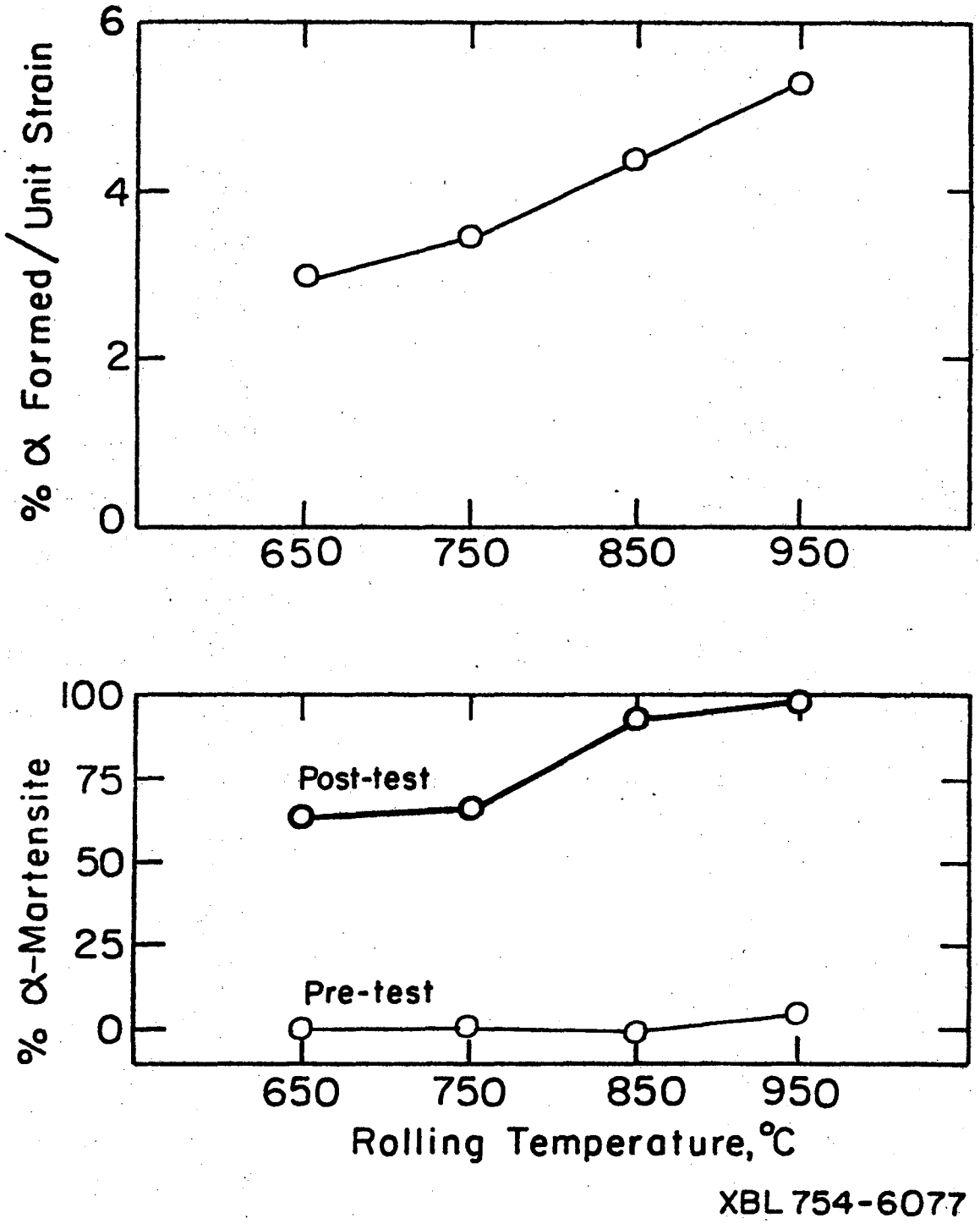

Fig. 10. γ to α transformation characteristics for 12-3 alloy.

temperatures) correlates with the increase in austenite stability -- with respect to athermal transformation -- found with increasing plastic deformation; both findings lend weight to the proposal that the formation of athermal α' (and possibly strain-induced α' as well) is controlled by the energy barrier to propagation through a deformed matrix.[40-43]

The 12-5 alloy (Fig. 11) exhibited approximately constant yield and ultimate strengths, although a slight decrease in strengths may be present as the rolling temperature increases. The property that sees the most change is the elongation; it almost doubles as the rolling temperature proceeds into the γ region (Fig. 12). This change in elongation is felt to be due to the increase in rate of formation of α' per unit strain as the rolling temperature is increased (Fig. 12).

The rate of formation of α' per unit strain has been demonstrated to be the controlling factor in determining a given TRIP steel specimen's elongation; either a too-low or too-high rate of formation will decrease elongation.[3, 44, 45] This optimization of α' formation rate appears to be present in the 12-5 alloy, as there is maximum in elongation at the 850C rolling temperature. Thus, at temperatures less than 850C the rate of formation is too low and at 950C the rate is too high, and uniform elongation decreases for either test condition.

The rate of α' formation for the rolling temperatures which exhibit high elongation is considerably less than that found for the high-elongation 12-3 alloy (Fig. 8). But, as the 12-5 alloy α' would be considerably higher in strength due to its increased carbon content, less of the 12-5 alloy α' would be needed to prevent necking and yield increased uniform elongation. The increase in the rate of α' formation during tensile plastic deformation is again, as in the 12-3 alloy, attributed to a decrease in matrix disorder.

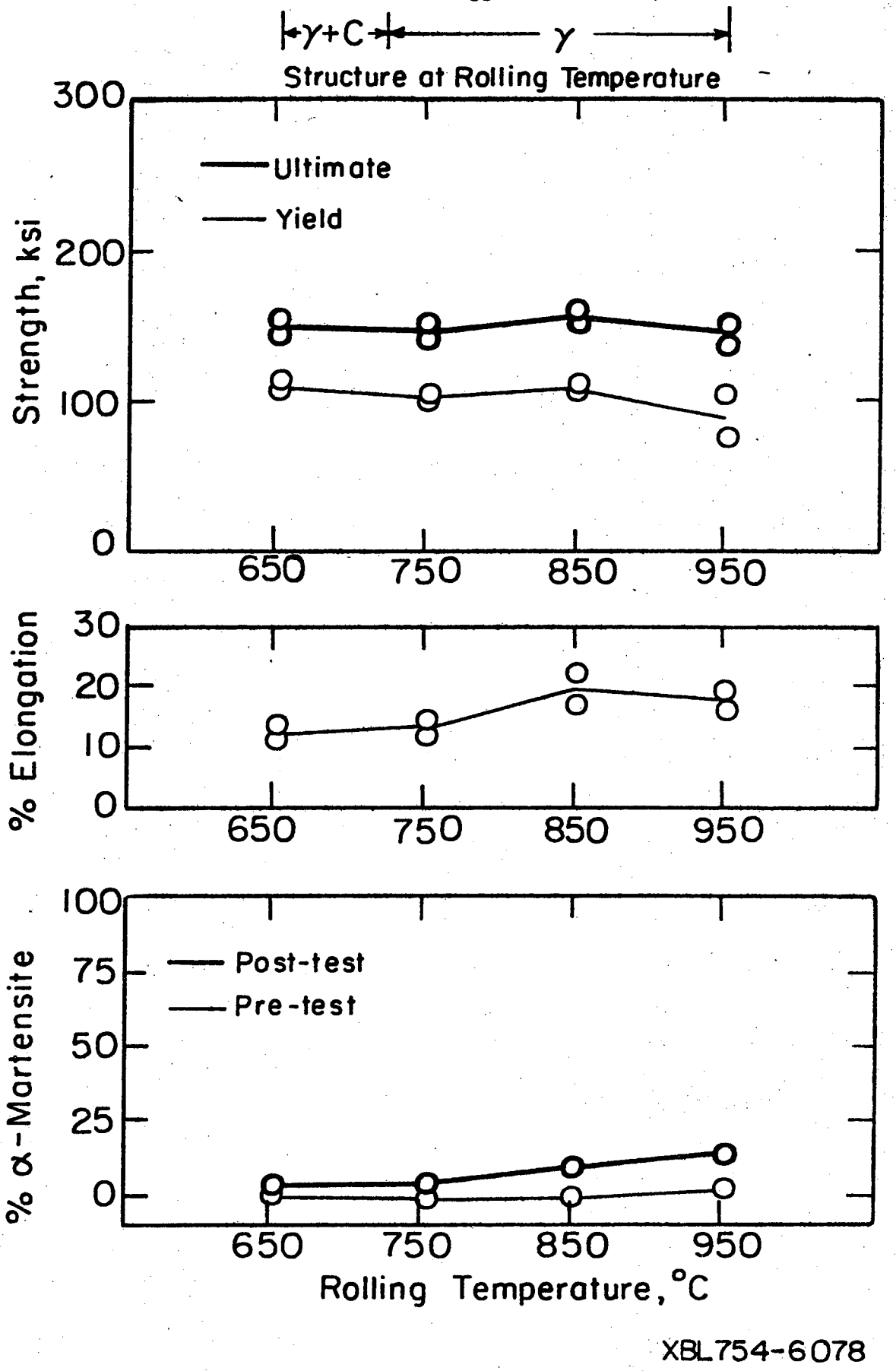

Fig. 11. Mechanical properties for 12-5 alloy.

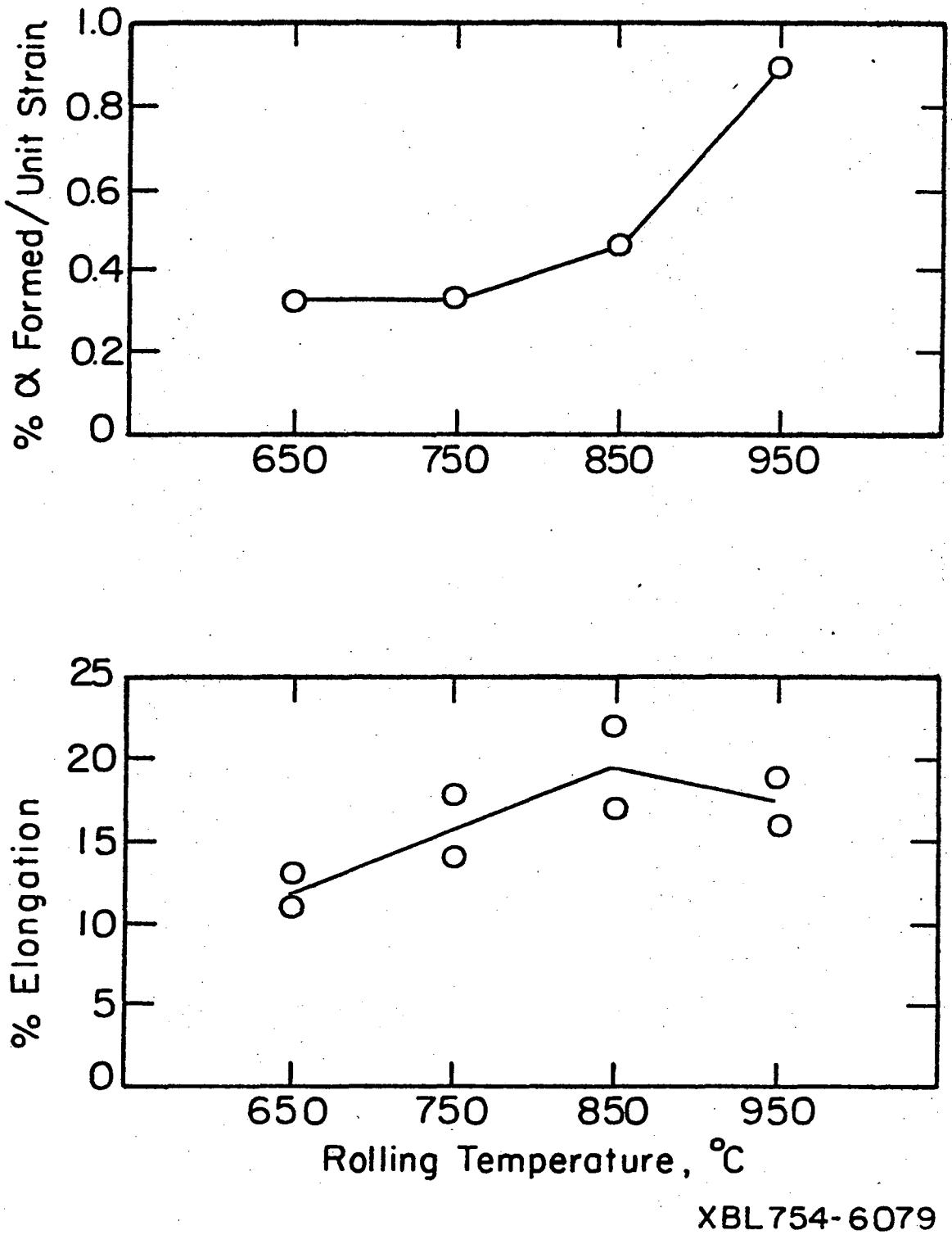

Fig. 12. γ to α transformation characteristics for 12-5 alloy.

The 12-8 alloy results, Fig. 13, span the regions of $\gamma+\alpha+C$, $\gamma+C$, and single phase γ. The ultimate strength is essentially constant and the yield strength is relatively unaffected by this change in phase stability regions, although the yield strength does exhibit a slight downward trend as one goes from region to region. There is, however, a decrease in strain hardening rate and a large change in elongation as the rolling temperature increases into the γ region. The lack of magnetic phase present after rolling in the $\gamma+\alpha+C$ region of the equilibrium phase diagram is consistent with the need for large α critical nuclei retarding nucleation in this region of the phase diagram.[22,23]

Both carbide precipitation and α' transformation kinetics are thought to influence the change in elongation found in this alloy. All the specimens tested were essentially retained austenite as-rolled, and transformed to approximately 5 percent α' during testing. This constant amount of α' formed per test, coupled with the increase in elongation with increasing rolling temperature, results in substantial changes in the average amount of α' formed per unit strain as a function of rolling temperature (from 0.67%/unit strain for the 650C test specimen to 0.16%/unit strain for the 950C test specimen). An increase in the carbon content of the matrix, as the rolling temperature increases from the $\gamma+\alpha+C$ region to the $\gamma+C$ region and then into the γ region, is postulated to be responsible for this decrease in strain-induced α' formation rate through increased stability of the γ matrix by the increased carbon content.

The α' formed in this 0.8% alloy would be exceedingly high strength. Thus, although this alloy exhibits approximately half the rate of α'

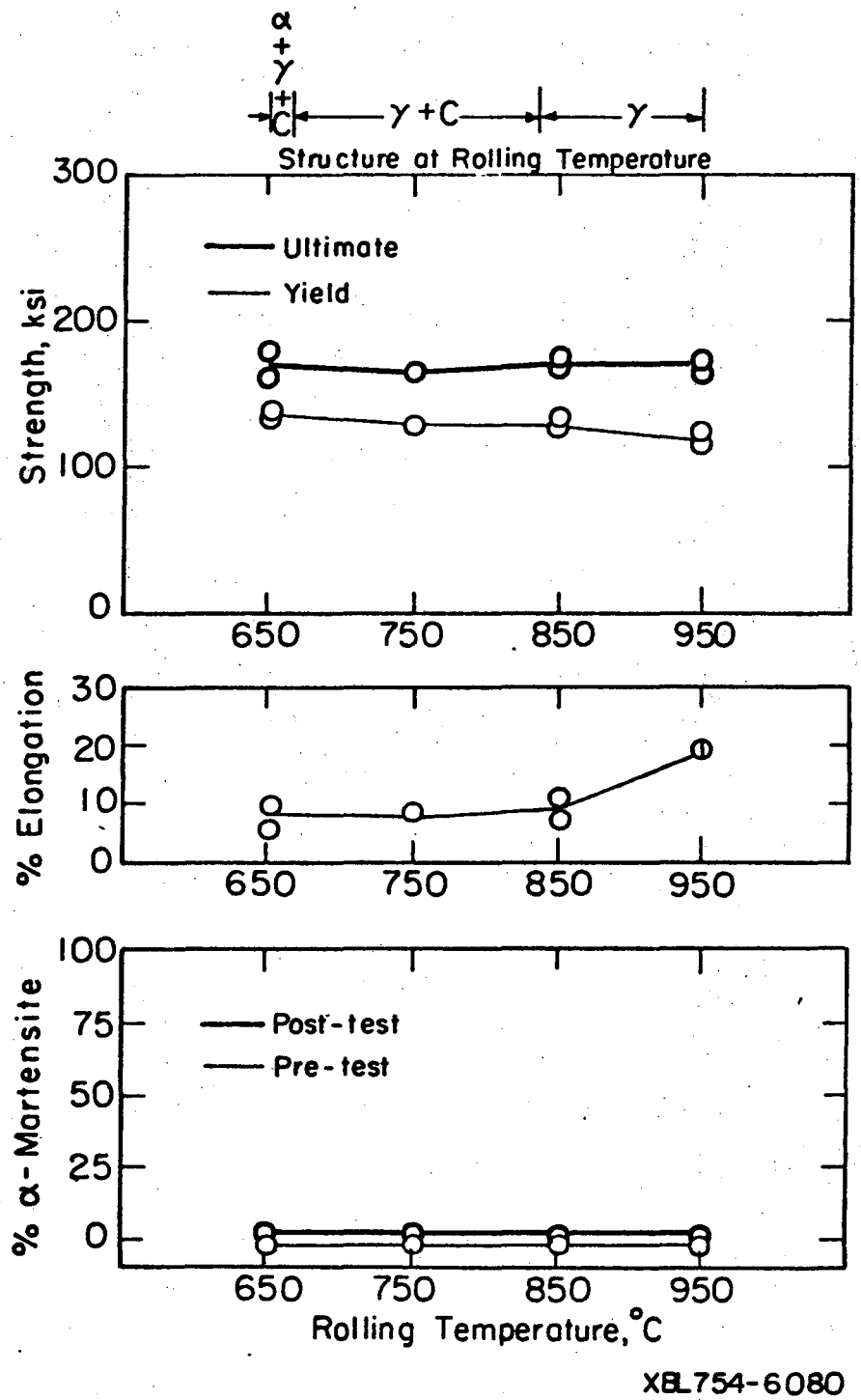

Fig. 13. Mechanical properties for 12-8 alloy.

formation in the high-elongation test as was found in the optimum elongation 12-5 alloy specimen, the 12-5 alloy discussion of α' formation rate controlling elongation may be applicable for this alloy also.

Carbide precipitation at grain boundaries may also influence elongation in this alloy. No carbide precipitation should take place prior to hot rolling, as the alloy specimens were austenitized and then immediately placed in the hot rolling furnace without an intermediate cooldown. Carbide clustering and/or precipitation is thermodynamically possible during hot rolling and during furnace soaks between rolling passes when the alloy is processed in either the γ+C or γ+C+α regions of the phase diagram. Grain boundary precipitation and/or clustering would be the first to take place, as grain boundaries are known to be preferred precipitation sites. This precipitation reaction was expected to be retarded by the grain boundary network being continually broken up both by plastic deformation during rolling and by plastic deformation-induced recrystallization taking place during the soak periods between rolling passes.[46] Nevertheless, grain boundary carbide films are known to have a profound effect on fracture properties, and are expected to limit elongation at the low processing temperatures.[47-49]

The 12-10 alloy yields results (Fig. 14) essentially the same as those found in the 12-8 alloy.

The composite results for the different alloys are replotted as a function of composition at a constant rolling temperature in Figs. 15-18. Once again the equilibrium phases at the given rolling temperature are indicated above the mechanical properties. All rolling temperatures exhibit essentially the same ultimate strength, yield strength and % α'

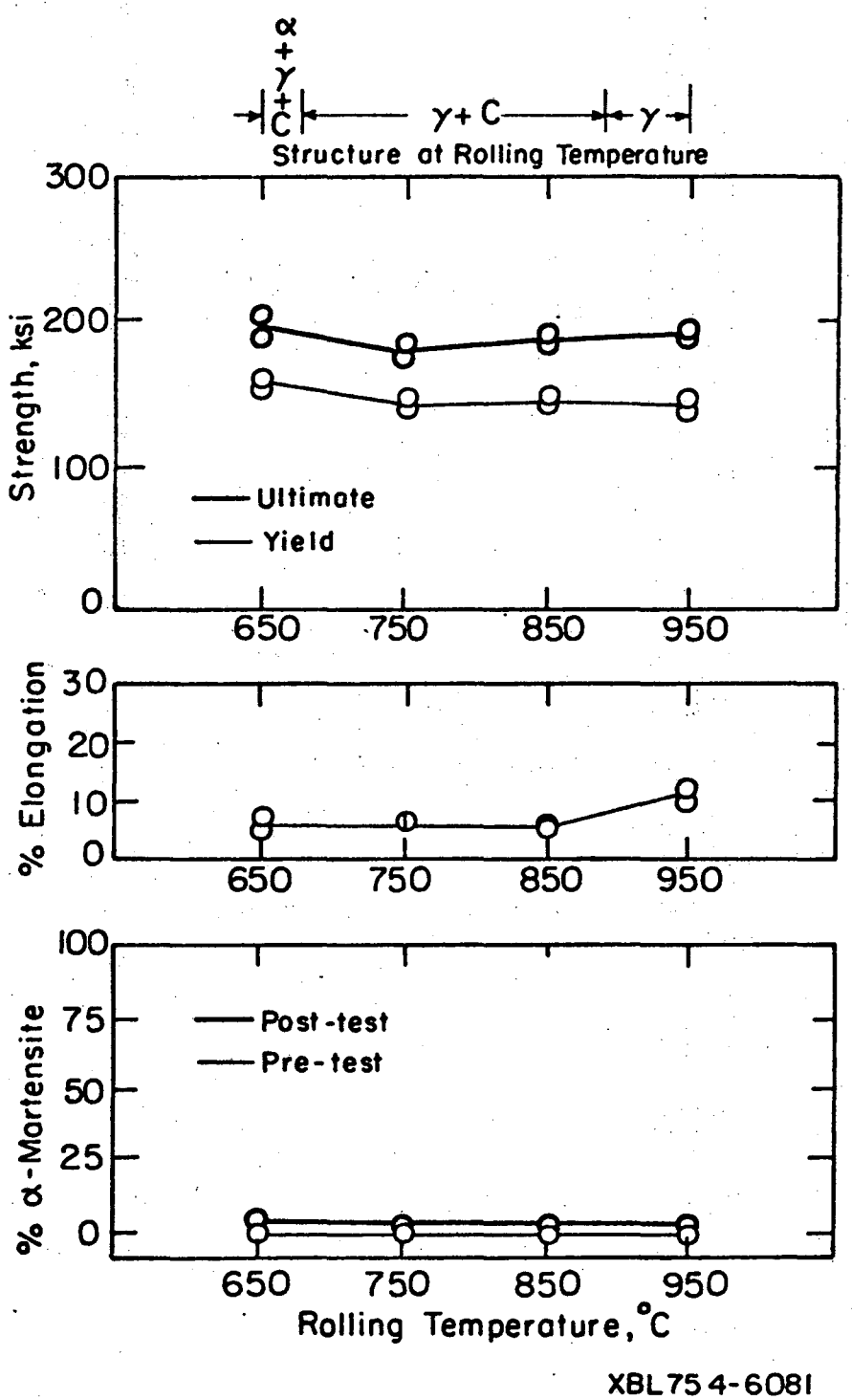

Fig. 14 Mechanical properties for 12-10 alloy.

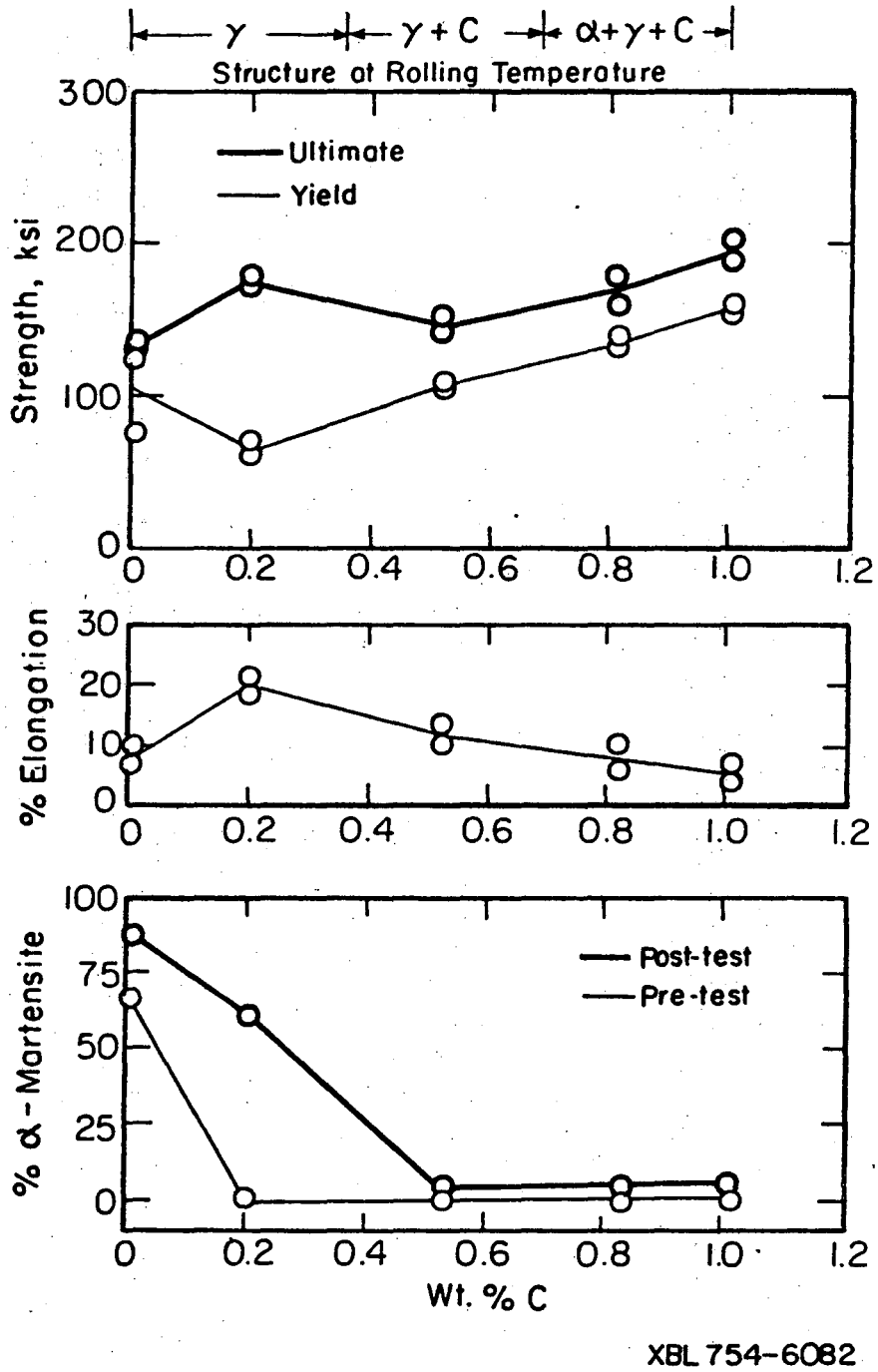

Fig. 15. Mechanical properties for 12-C alloys rollec at 650C.

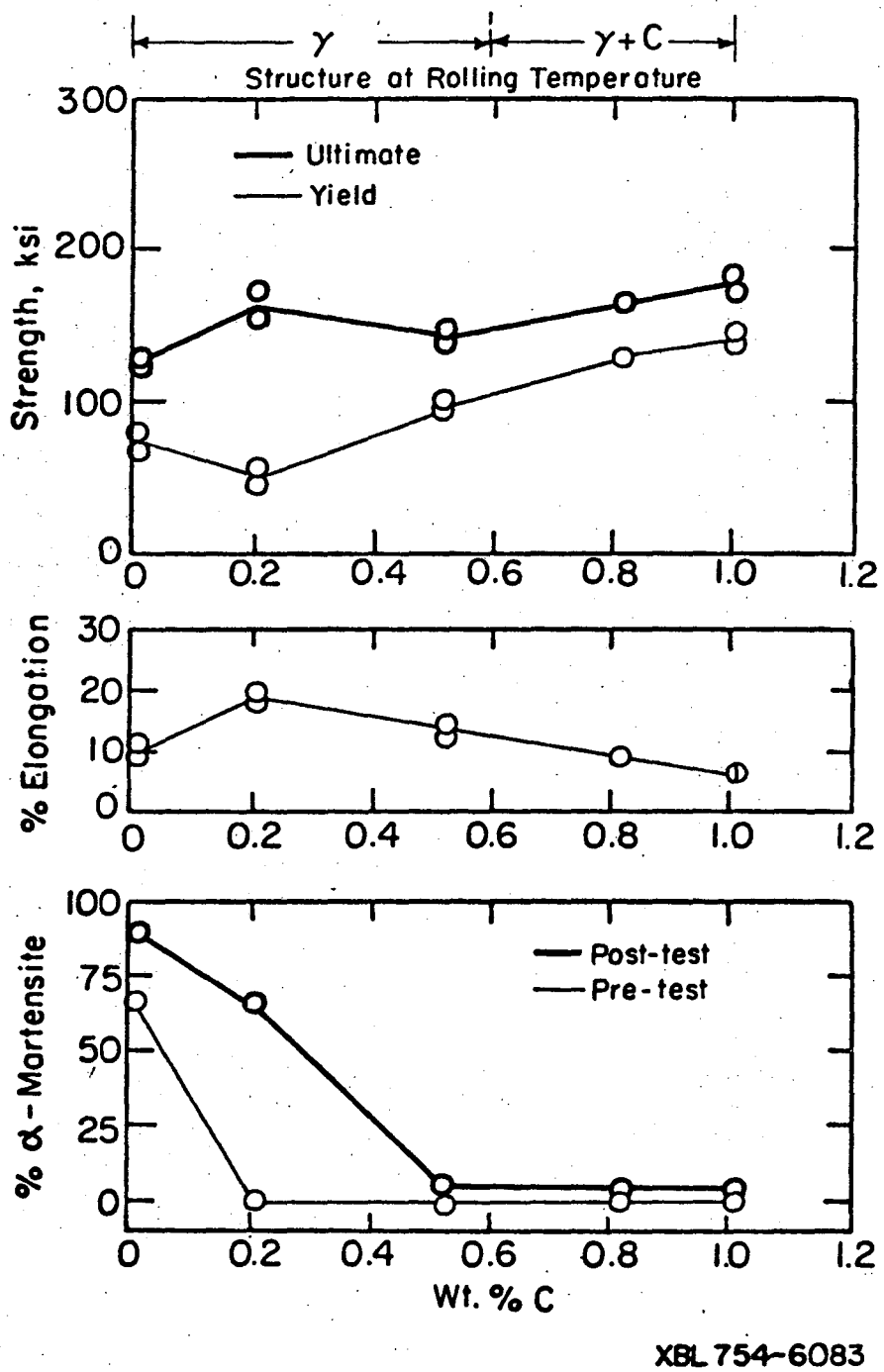

Fig. 16. Mechanical properties for 12-C alloys rolled at 750C.

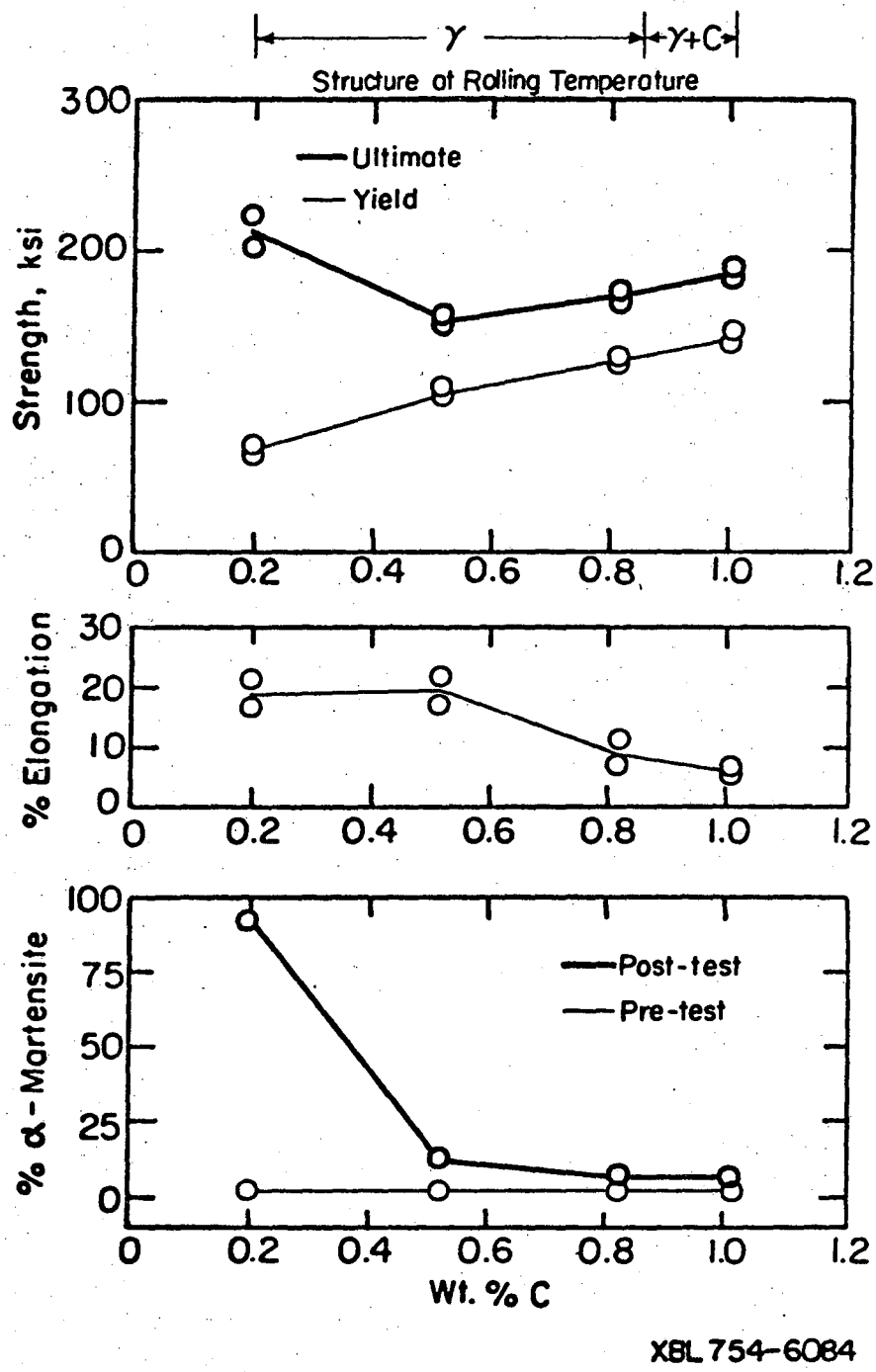

Fig. 17. Mechanical properties for 12-C alloys rolled at 850C.

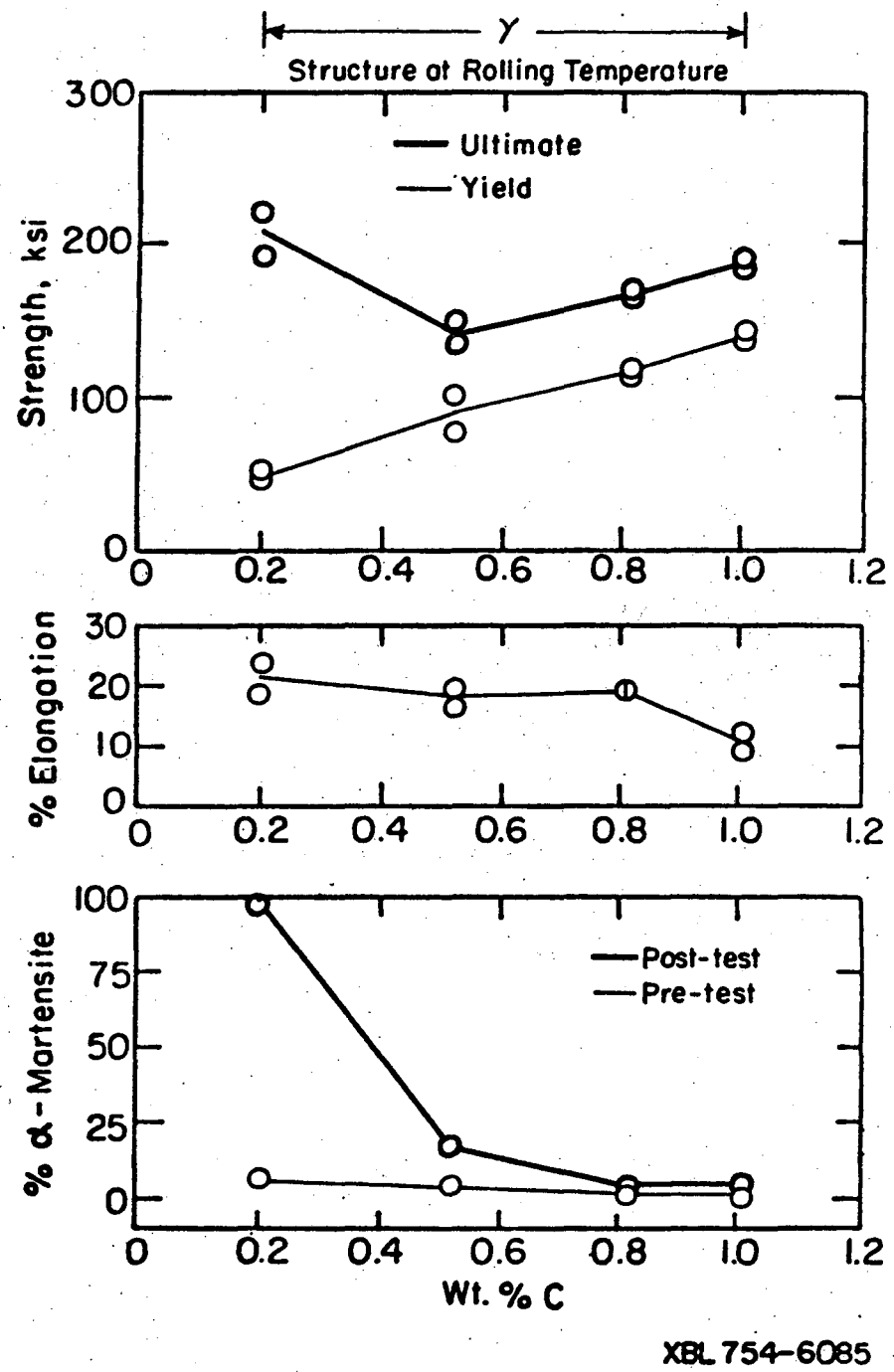

Fig. 18. Mechanical properties for 12-C alloys rollec at 950C.

characteristics. The 12-0 alloy is close to 75% α' as-rolled and exhibits a strain-induced transformation during testing. The rest of the alloys are essentially retained γ as-rolled. The 12-3 alloy exhibits a considerable amount of strain-induced transformation; the 12-5 exhibits a slight amount, and the other alloys are close to remaining retained γ throughout testing.

In general, it can be seen that the yield strengths increase as the carbon percentage increases. The exception to this is the decrease in yield strength going from 0 to 0.2% C; this is due to the 12-0 alloy's containing considerable α' as-rolled while the 12-3 alloy was essentially retained γ. The elongation for the metastable γ alloys is greatest in alloys treated in the γ region, and decreases as one progresses to the γ+C and then to the $\gamma+\alpha$+C region. Elongations for a given rolling temperature are high if an alloy is rolled in the γ range. Alloys rolled within the γ range exhibit elongations of approximately the same magnitude even with widely varying yield strengths; this is praticularly evident for the alloys rolled at 950C, where the yield strength is more than doubled with very little change in the total elongation. As rolling temperature is increased the yield strengths of a given alloy decrease, and their respective elongations tend to increase.

2. <u>Fe-10% Mn Alloys</u>

The carbon content and as-solution-treated structures of the 10% Mn alloys basically follow those of the 12% Mn alloys. The as-solution-treated structure of the 10-3 alloy fell in the $\alpha'+\varepsilon'+\gamma$ region, the 10-5 fell in the $\alpha'+\gamma$, and the 10-8 and 10-10 alloys fell in the γ region.

The mechanical properties for the 10-3 and 10-5 alloys are plotted in Figs. 19 and 20 as a function of rolling temperature. Both compositions

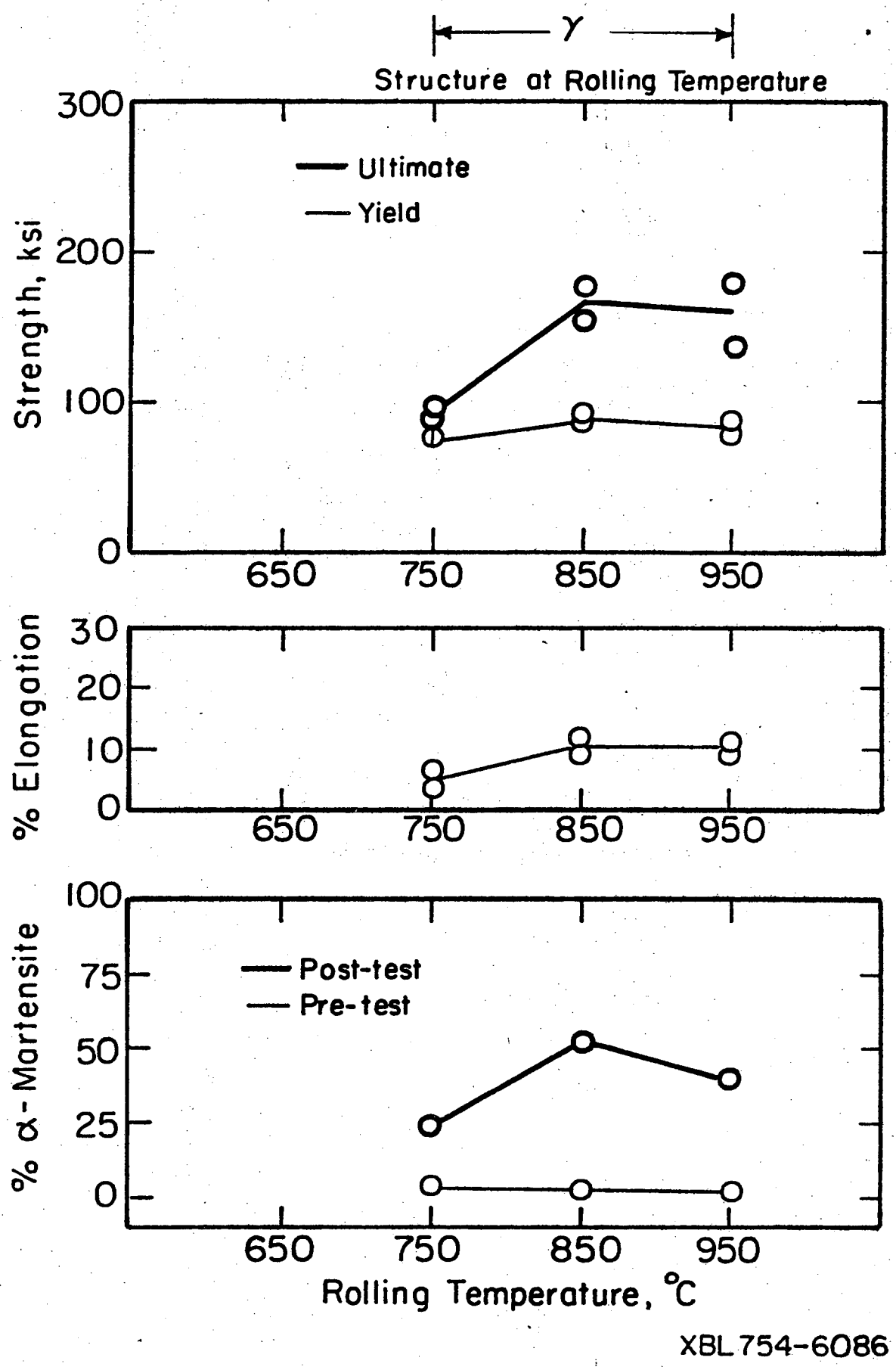

Fig. 19. Mechanical properties for 10-3 alloy.

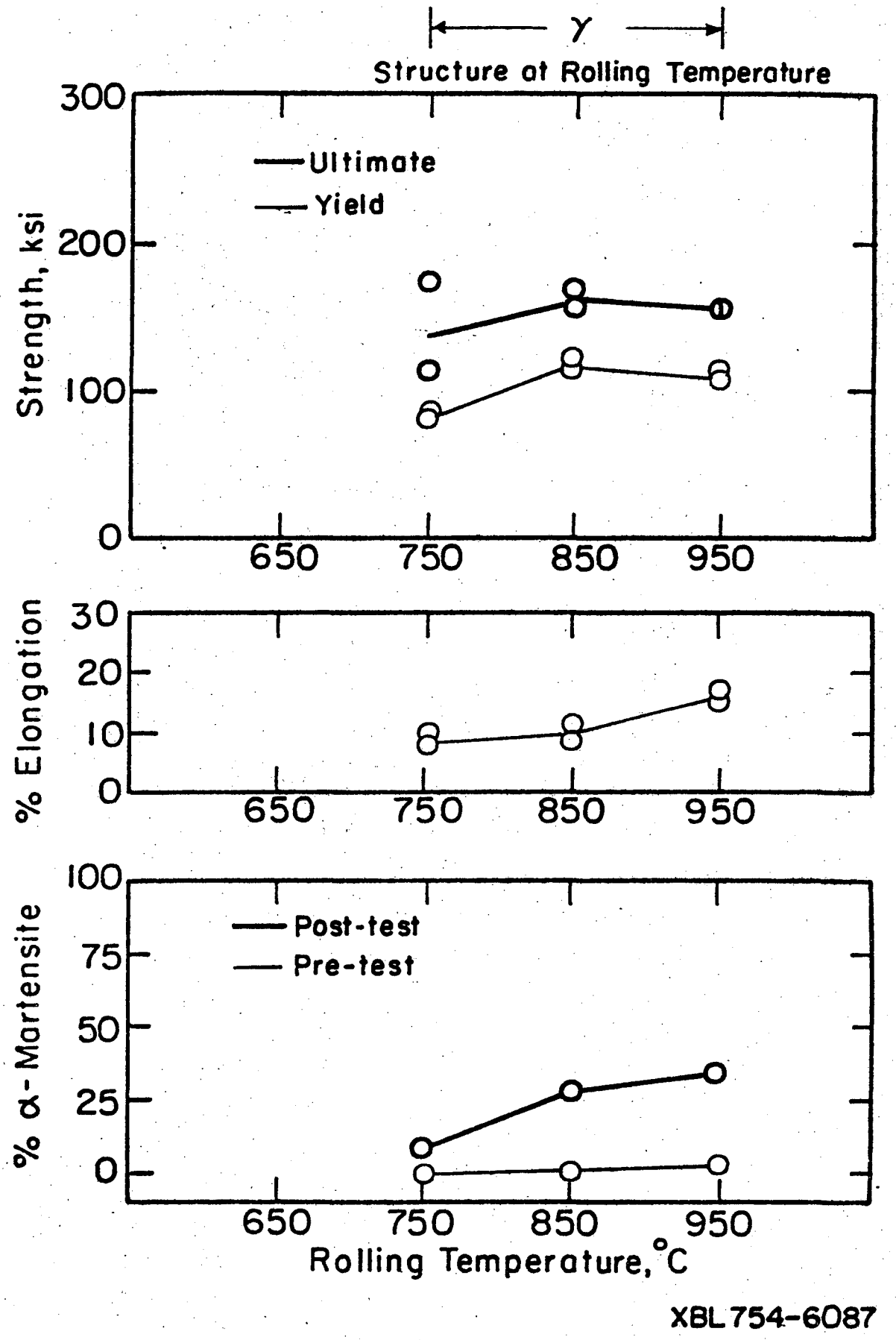

Fig. 20. Mechanical properties for 10-5 alloy.

are essentially retained γ as-rolled for all test temperatures and exhibit a γ to α' phase tranformation during straining.

The amount of post-test α' for these alloys tends to increase as the rolling temperature increases; however, so does the elongation. Nevertheless, the average rate of α' formation per unit strain does tend to increase as the rolling temperature increases. This correlation between increasing formation rate and increasing elongation again illustrates the control of transformation rate over elongation, as discussed in the 12% Mn alloys.

The mechanical property changes as a function of carbon content for the 850C and 950C rolling temperatures are similar and are given in Figs. 21 and 22. All alloys are essentially retained γ as-rolled, with the γ stability during testing increasing as the carbon content increases. This is demonstrated for the 950C test specimens by the decrease in the α' formation rate and in the post-test α' (Fig. 23). The yield strength tends to increase and the elongation tends to decrease as the carbon content increases.

The yield and ultimate strengths for a given carbon content are relatively unaffected by a change in rolling temperature, except for the 10-10 alloy. For this alloy the strengths at the 850C rolling temperature are greater than those at 950C. This change in strengths would be due to carbide clustering and/or precipitation at dislocations in the γ matrix, which is to be expected at the 850C rolling temperature, as it falls into the γ+C region; the 950C rolling temperature is in the solely γ region.

3. Fe-8% Mn

The as-solution-treated structures of the 8% Mn alloys consisted of $\alpha'+\gamma$ for the 8-3 and 8-5 alloys, and γ for the 8-8 and 8-10 alloys. The mechanical properties for the 8-5 alloy are presented in Fig. 24 as a

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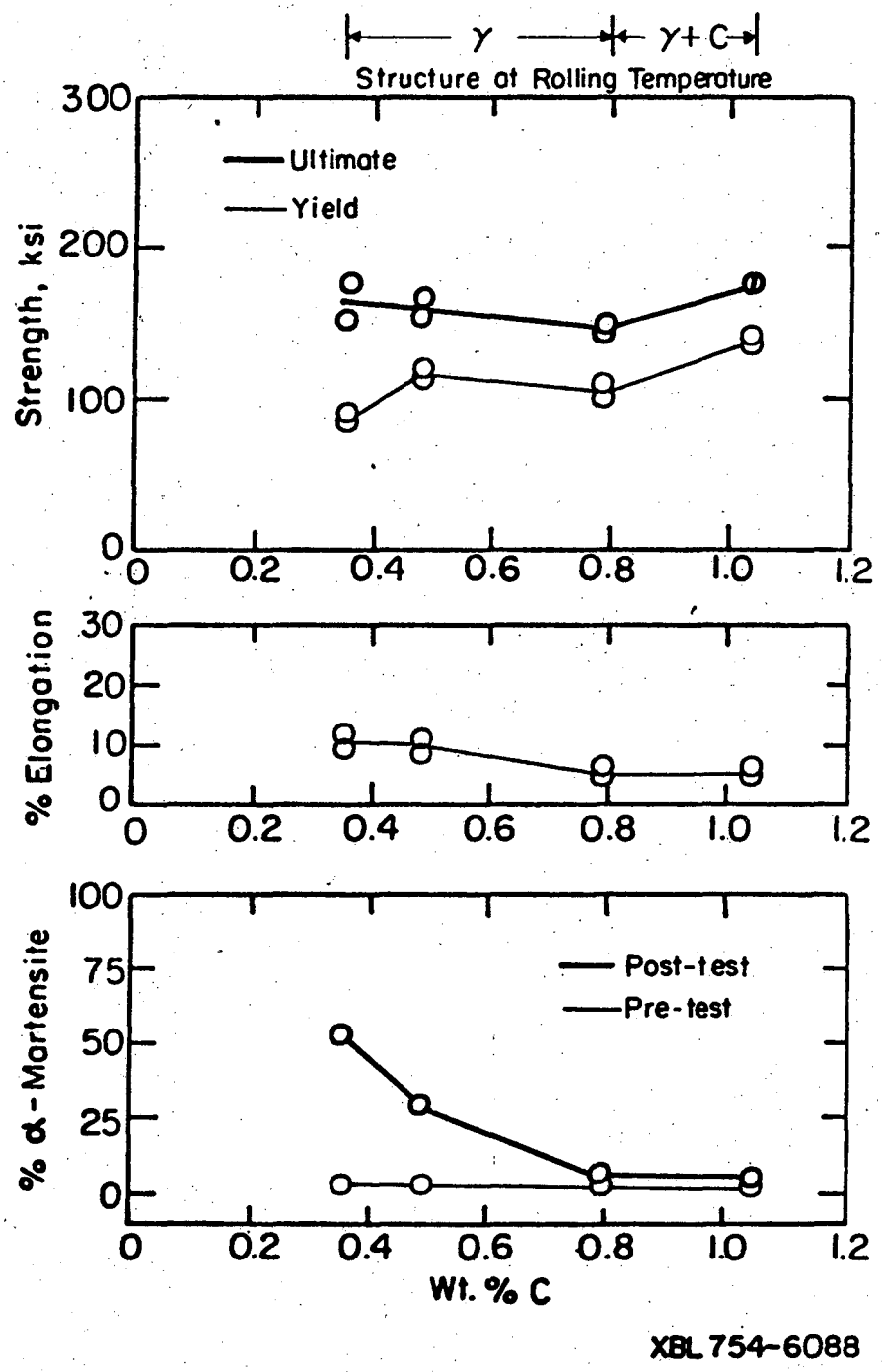

Fig. 21. Mechanical properties for 10-C alloys rolled at 850C.

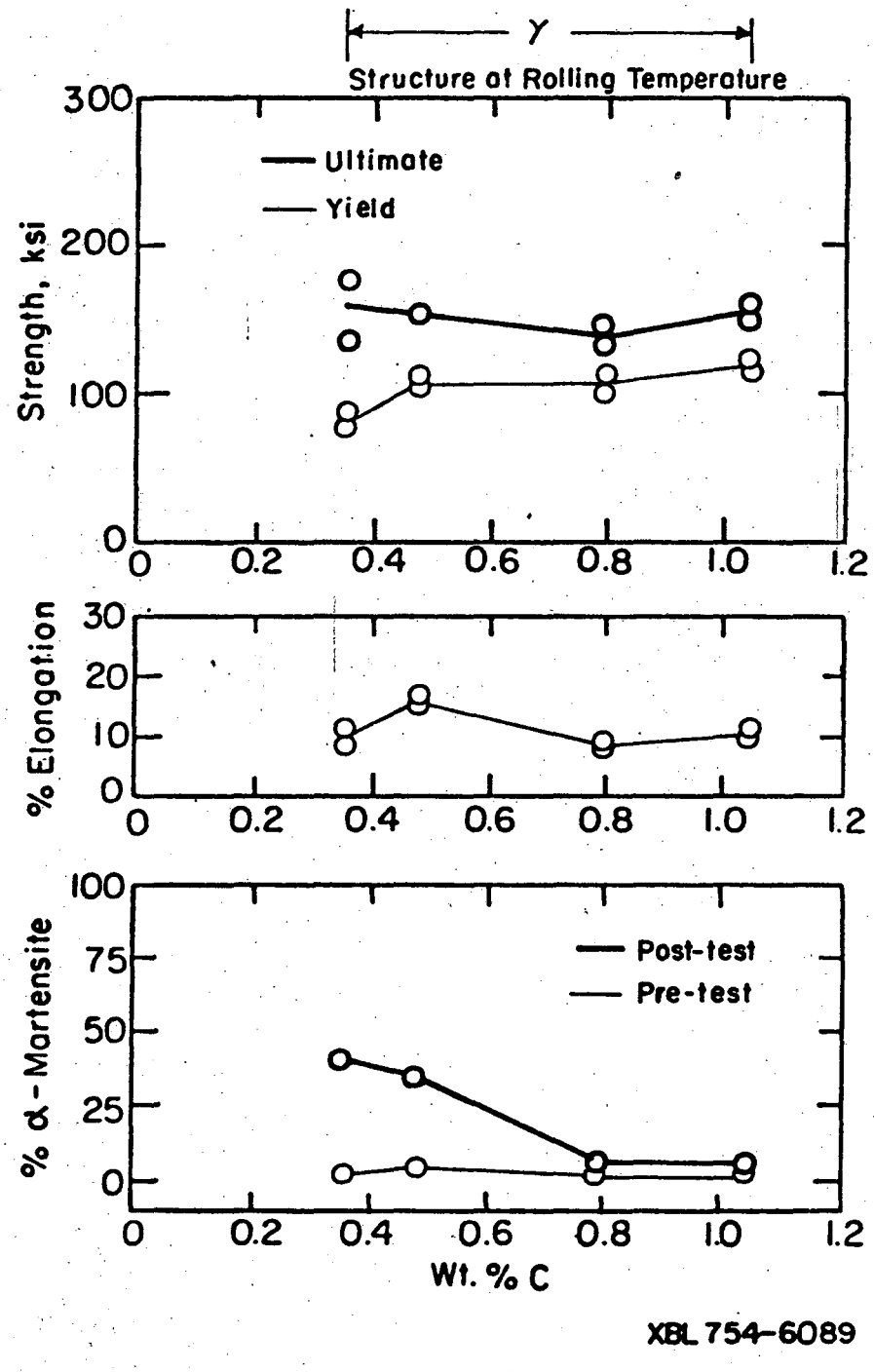

Fig. 22. Mechanical properties for 10-C alloys rolled at 950C.

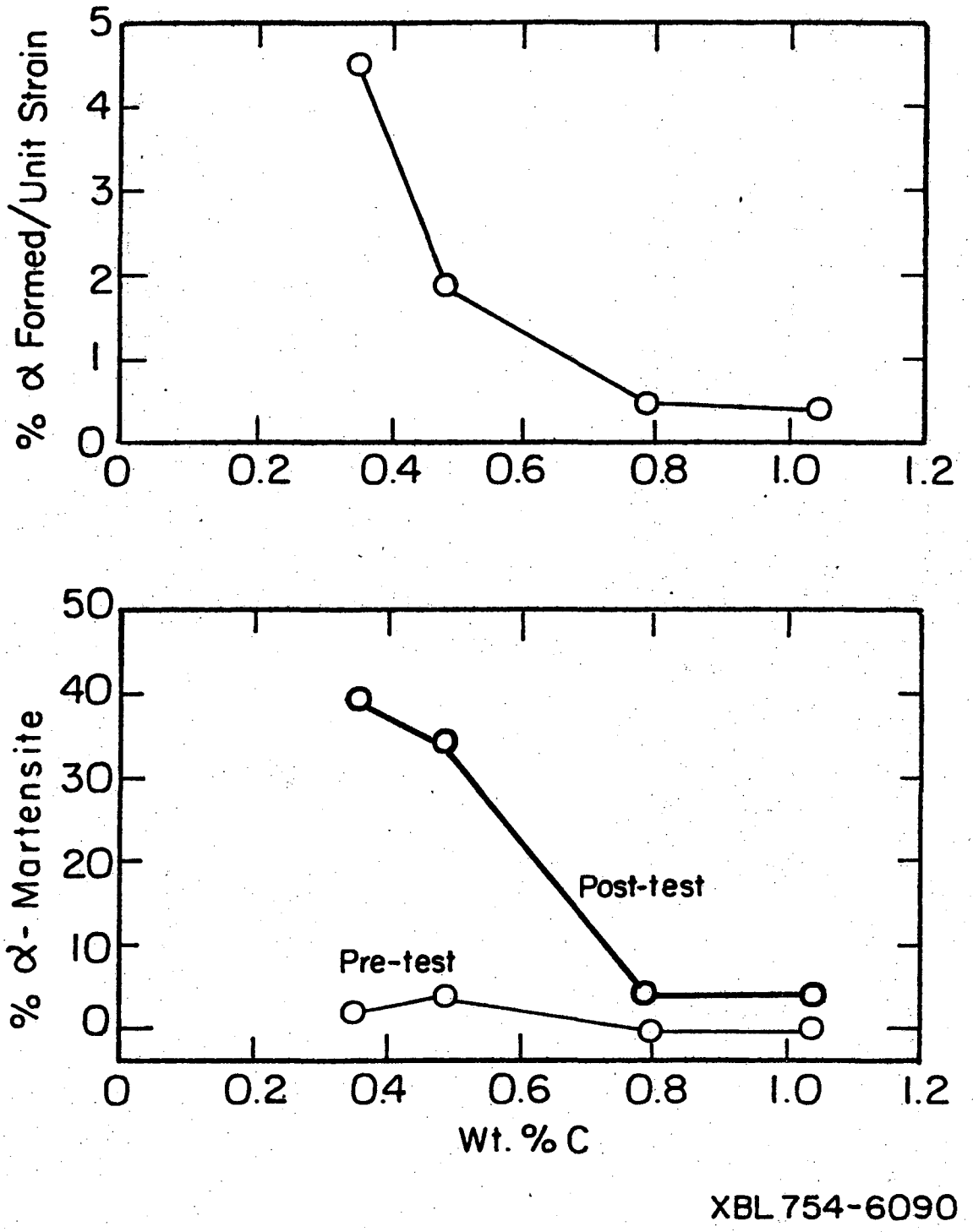

Fig. 23. γ to α transformation characteristics for 10-X alloy rolled at 950°C.

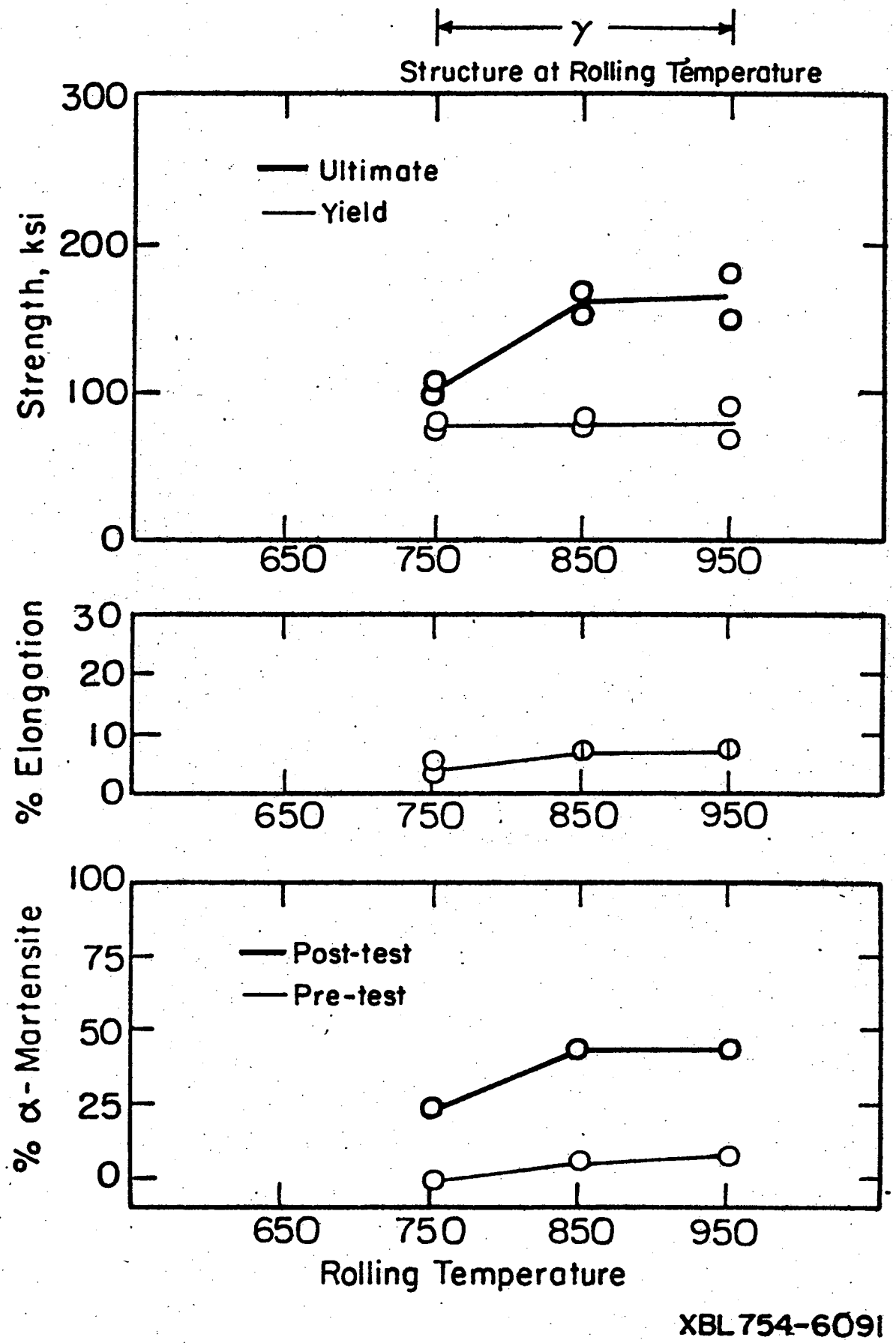

Fig. 24. Mechanical properties for 8-5 alloy.

function of rolling temperature, while those for all alloys rolled at 850C and 950C are presented as a function of carbon concentration in Figs. 25 and 26.

The 8-5 alloy properties (Fig. 24) show low elongations and moderate yield strengths. Elongation and percent pre- and post-test α' increase with increasing rolling temperature; all rolling temperatures were in the γ region.

The results show that, as one increases the rolling temperature and proceeds further into the γ region, the M_s temperature increases; this is indicated from the increase in α' present as-rolled.

The rate of α' formed per unit strain is very great for the 750C specimen (24%/unit strain), decreases to 5.4%/unit strain for the 850C specimen, and then increases for the 950C specimen (to 5.8%/unit strain). The plastic deformation retained in the 750C specimens resulted in an essentially retained γ matrix which was very unstable. Small amounts of strain induced a large amount of transformation to α'. The reduced retained plastic deformation in the 850C specimens allowed a rise in the M_s temperature and a resultant mixed room-temperature structure of athermal α' and γ. The resistance to strain-induced transformation was less for the 850C specimens than for the 750C specimens due to the reduced plastic deformation of the matrix, but the thermodynamic driving force was also less due to the formation of the athermal α'. Thus, the α' formation rate for the 850C specimens was less than that for the 750C specimens. The 950C specimens exhibited a slight increase in both athermal α' and strain-induced transformation rate, with respect to the 850C specimen -- as would be

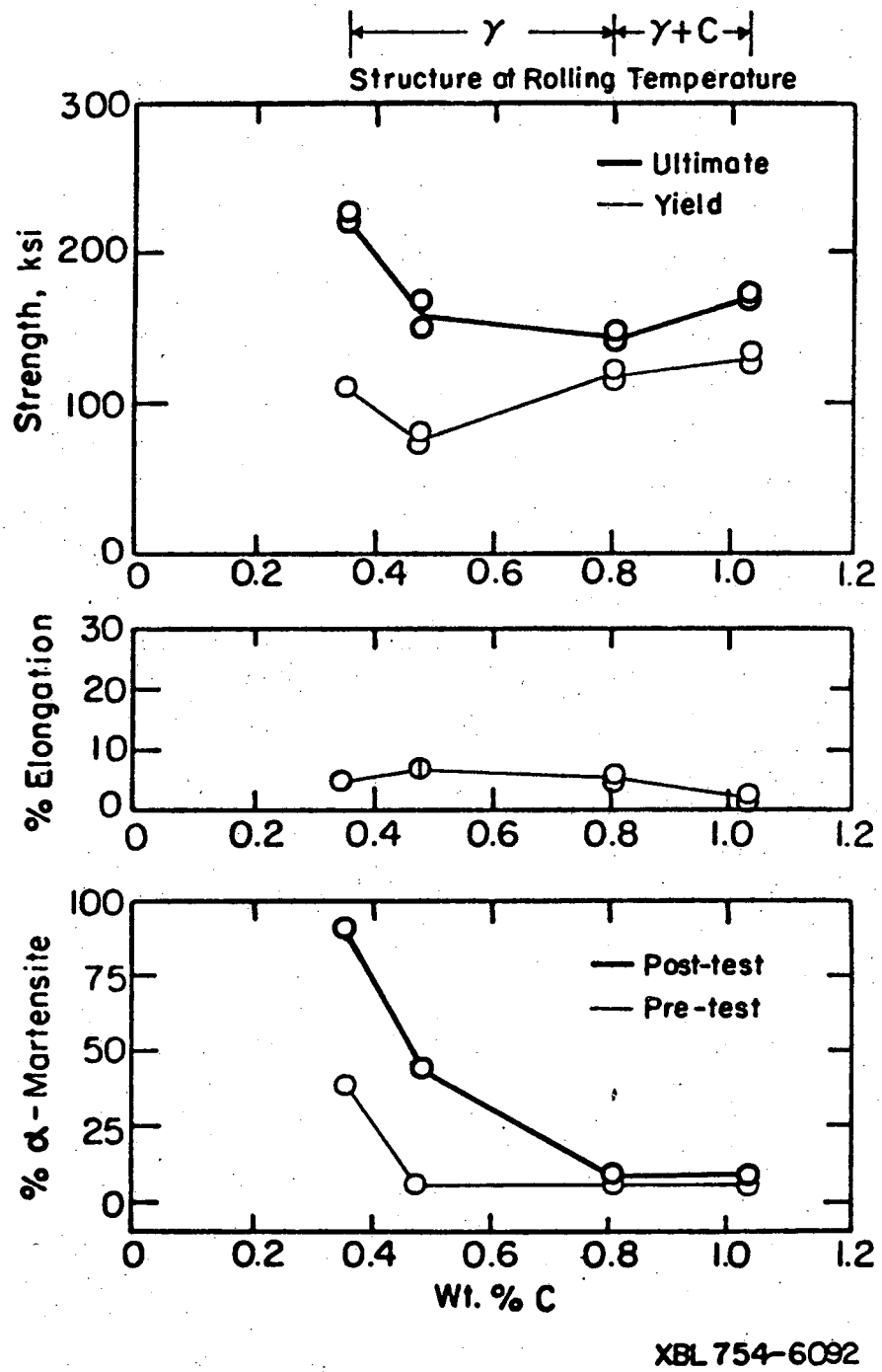

Fig. 25. Mechanical properties for 8-C alloys rolled at 850C.

0 0 0 0 4 2 0 8 1 4 2

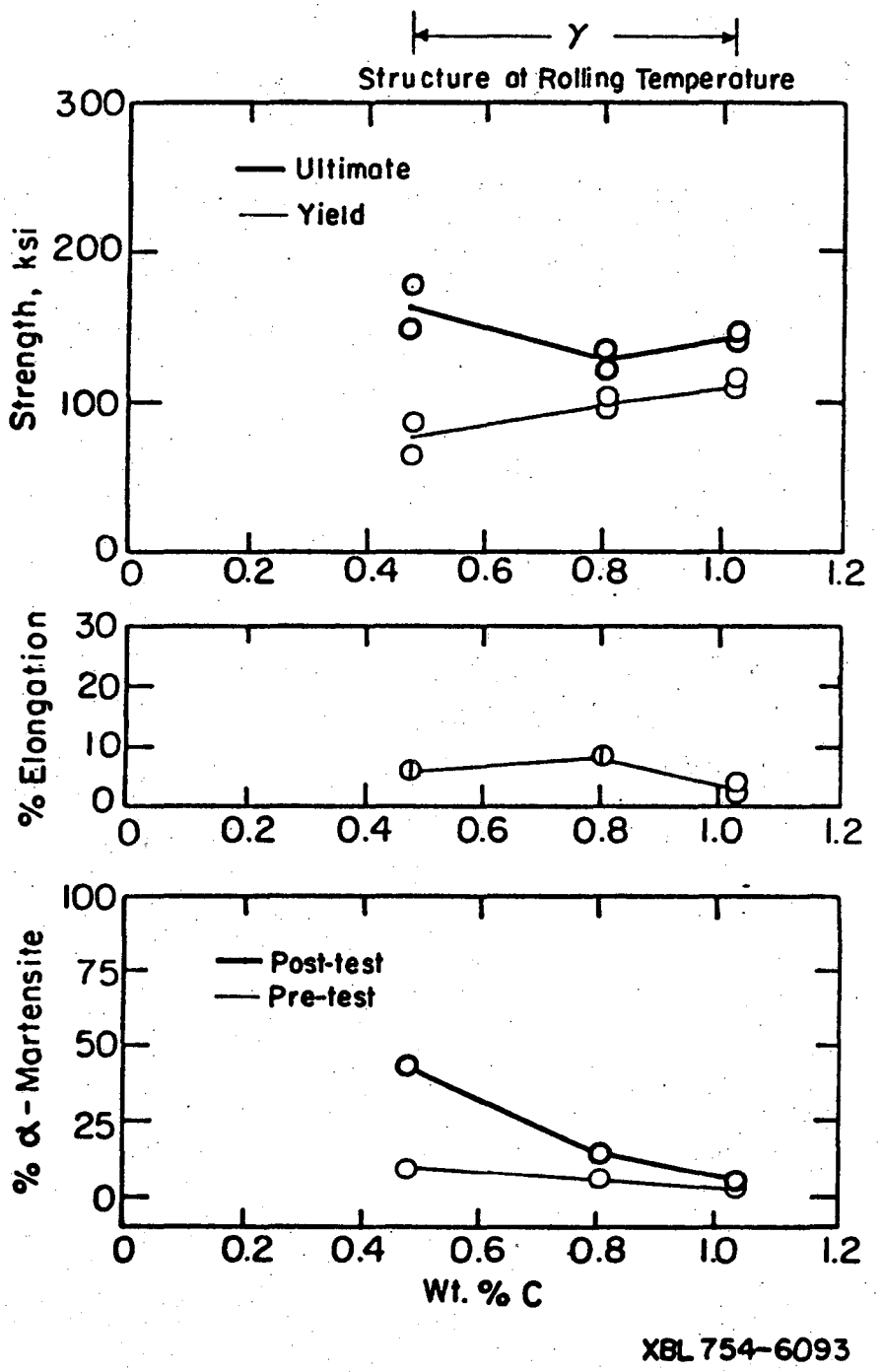

Fig. 26. Mechanical properties for 8-C alloys rolled at 950C.

expected due to the decrease in retained plastic deformation decreasing the γ matrix stability.

The 850C and 950C results (Figs. 25 and 26, respectively), show increased yield strength and γ stability and decreased elongation as carbon content is increased. The exception to these general conclusions is the 8-3 alloy at 850C. This alloy exhibits an uncharacteristically high yield strength due to the 38% α' present as-rolled.

4. Fe-6% Mn

Only three 6% Mn alloys were processed. The as-solution-treated 6-5 and 6-8 alloys were in the $\alpha+\gamma$ region, while the 6-10 alloy was in the γ region. These alloys were processed only at 850C and 950C; the results are given in Figs. 27 and 28.

The yield strength for both rolling temperatures was found to vary in a similar manner as a function of carbon content. It first decreased as the carbon content increased, due to the decrease in as-rolled α' with increasing carbon content; and then increased, due to increased carbon clustering and/or precipitation at γ dislocations, as the carbon content was further increased. The elongation of this alloy series was consistently low.

5. Fe-Constant % C Alloys

The data given in the previous sections were replotted for this section to yield changes in properties as a function of constant carbon content and changing manganese content. The 750C rolling temperature results are given in Figs. 29 and 30 and illustrate the effect of changing the amount of α' formed per unit strain and the problem of generalizing alloying element effects.

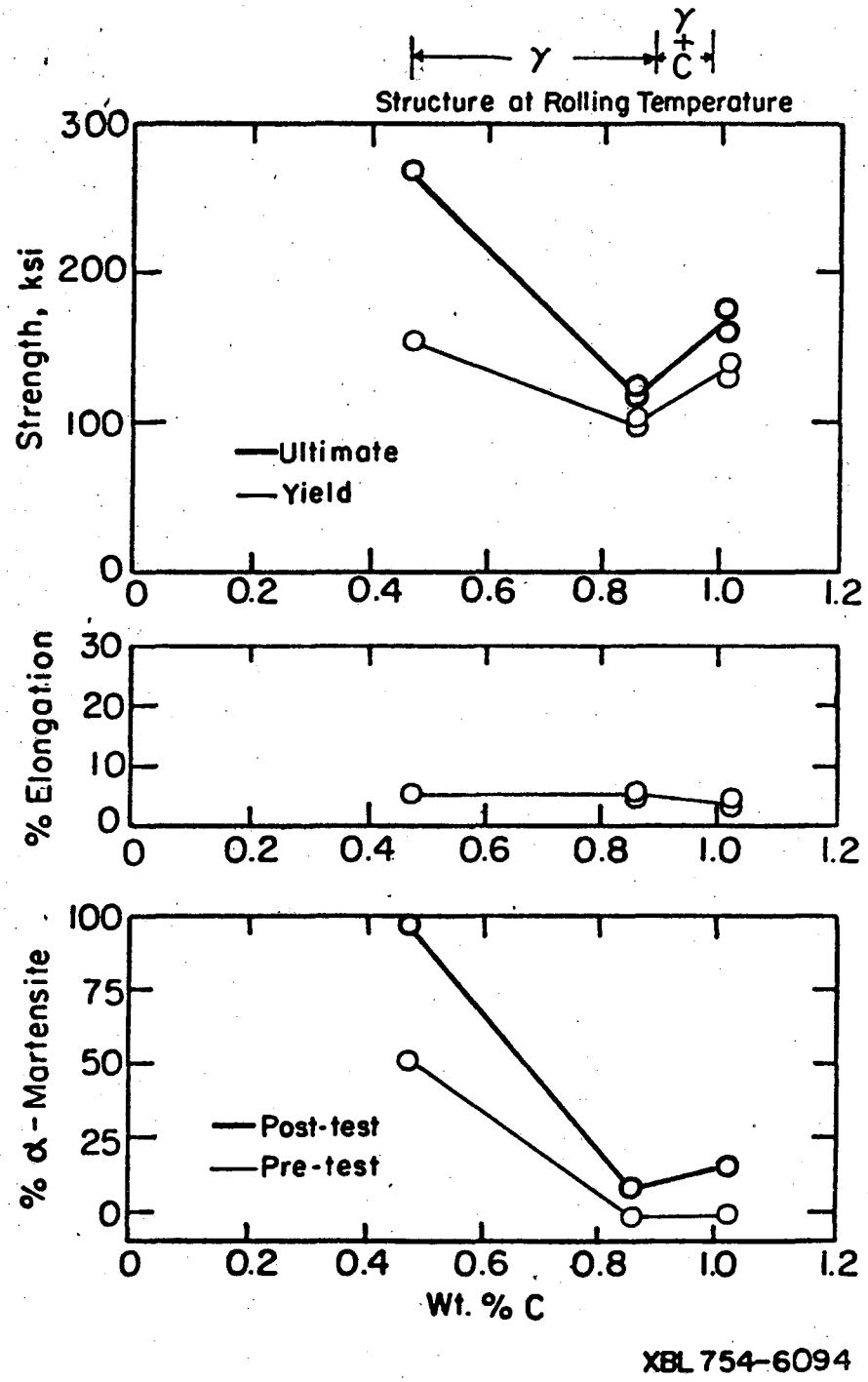

Fig. 27. Mechanical properties for 6-C alloys rollec at 850C.

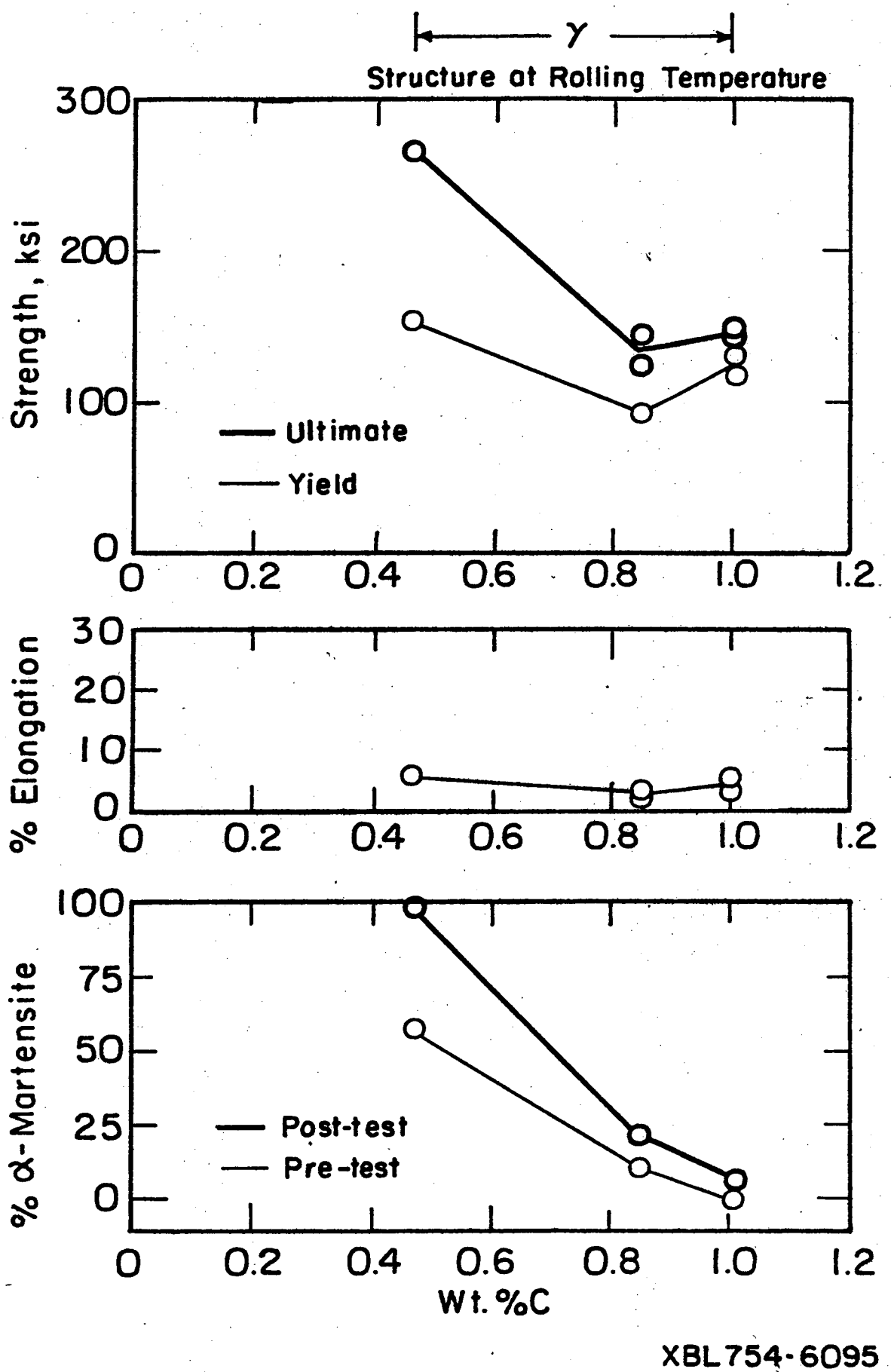

Fig. 28. Mechanical properties for 6-C alloys rolled at 950°C.

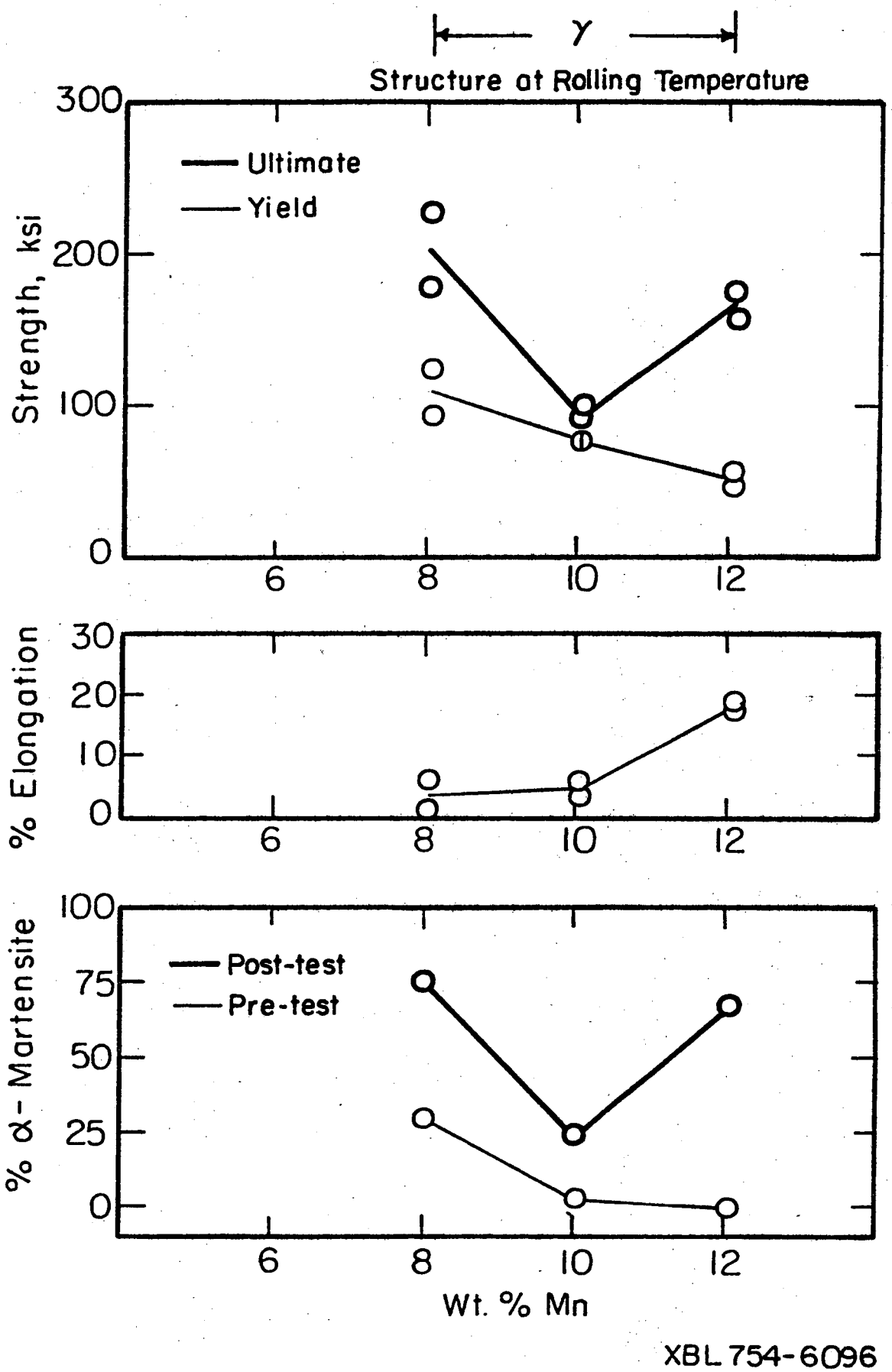

Fig. 29. Mechanical properties for Mn-3 alloys rolled at 750°C.

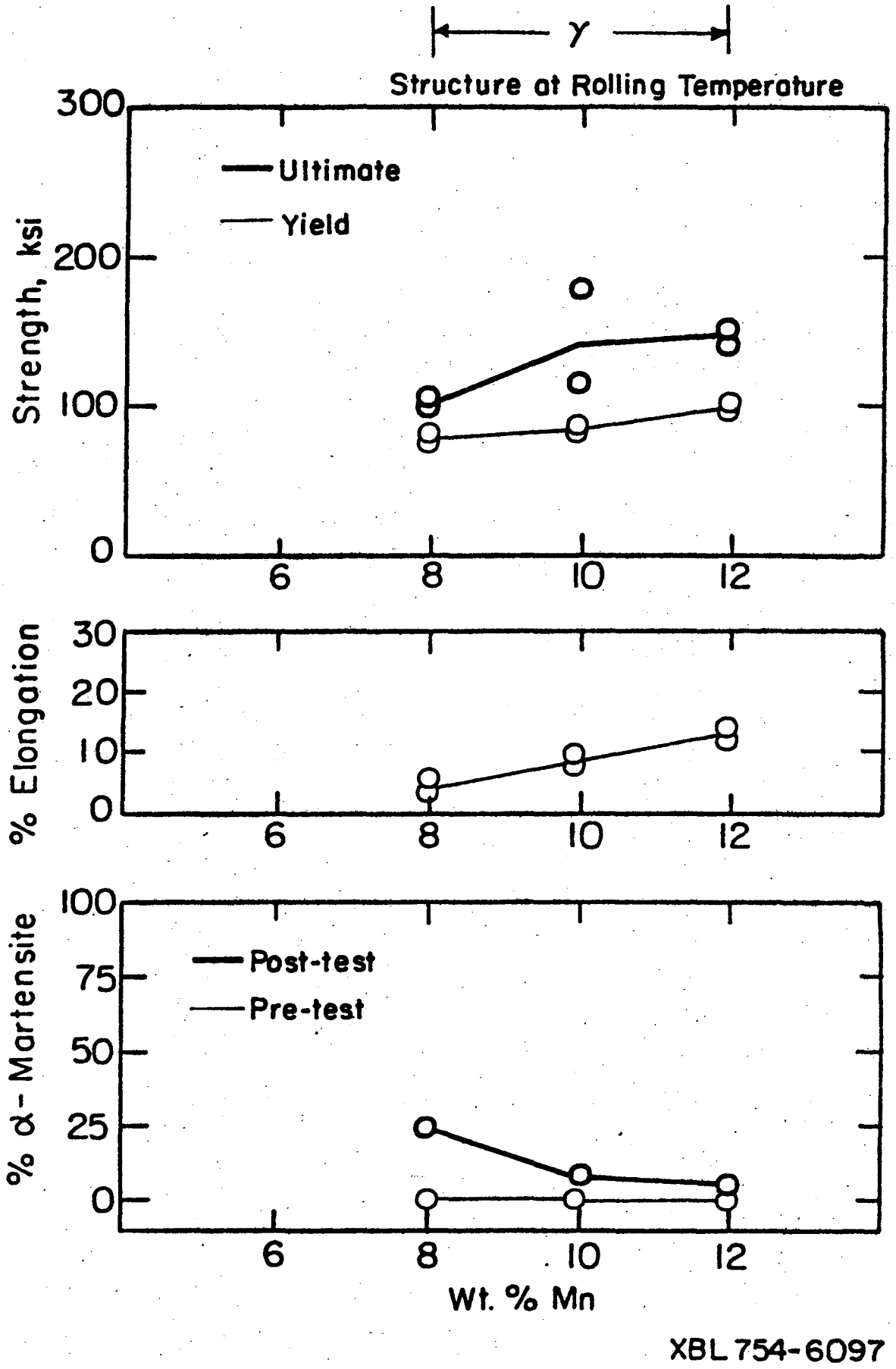

Fig. 30. Mechanical properties for Mn-5 alloys rolled at 750°C.

The Mn-3 alloys illustrate the effect on elongation of changing the amount of α' formed per unit strain. The elongation increases as the average amount of α' formed per unit strain decreases (Fig. 31) because the increase in Mn stabilizes the γ. The ultimate strength for this alloy series varies directly with the total amount of α' present at failure; the decrease in yield strength as the % Mn increases is caused by the decrease in as-rolled α'.

The Mn-5 alloy series also exhibits increasing elongation with decreasing amount of α' formed per unit strain (Fig. 32; note change in formation rate scale from Fig. 31), but the amount formed per unit strain needed to give high elongation is considerably below that of the Mn-3 alloy results. This can be due either to the increased strengthening received per unit of 0.5%C α' due to its increased carbon content, and/or to the formation of either narrow stacking faults or ε', both of which are favored by increasing Mn content.[50-54] The ultimate strengths for the Mn-5 alloys are not related to the percentage of α' present; in fact the ultimate strength increases as the percentage α' present at fracture decreases. This lends weight to the assumption of the presence of ε' as a second strain-hardening mechanism.

The as-rolled structure for all specimens is essentially retained γ and exhibits an increasing yield strength as the % Mn increases. This increase in strength can be explained as a combination of solid solution hardening due to increasing Mn content and an increase in carbon clustering or precipitation at γ matrix dislocations due to decreased randomness of the carbon distribution as the processing temperature approaches the γ to γ+C transformation temperature with increasing Mn content.

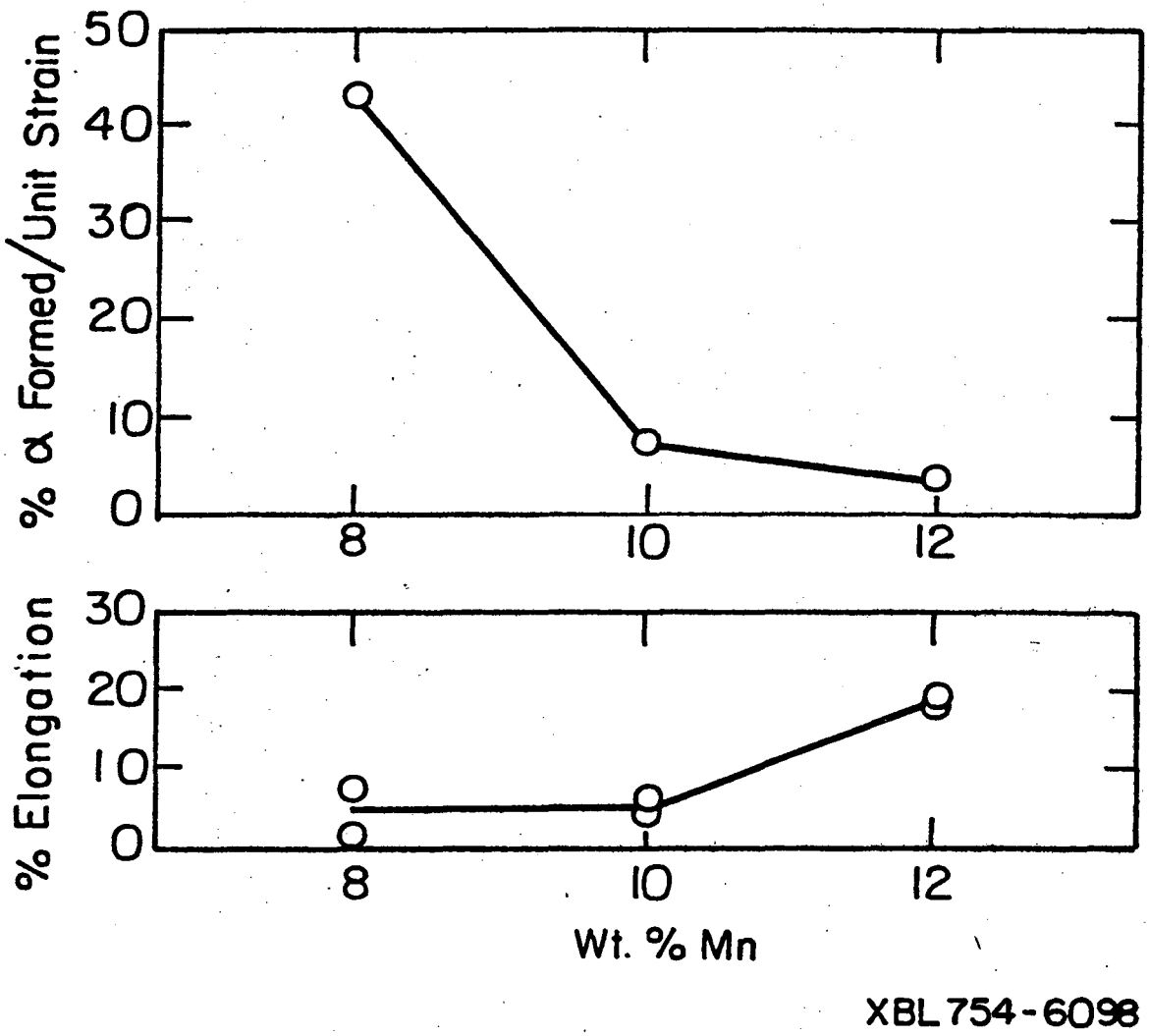

Fig. 31. γ to α transformation characteristic for Mn-3 alloy rolled at 750°C.

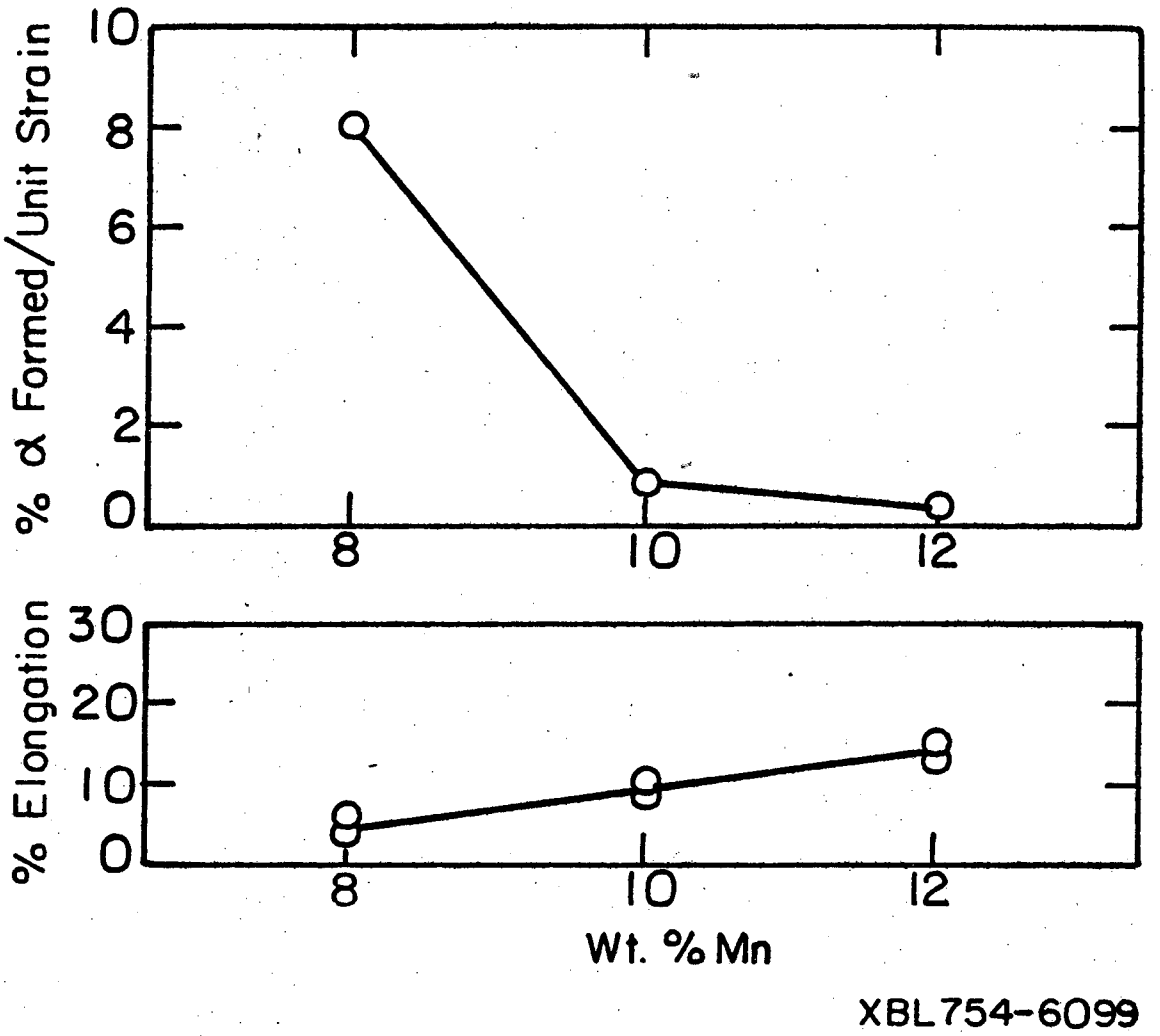

Fig. 32. γ to α transformation characteristics for Mn-5 alloys rolled at 750°C.

The changes in yield strength with increasing Mn content in Figs. 31 and 32 illustrate the difficulty in generalizing the effects of alloying additions. Increasing Mn in the Mn-3 alloy series decreases the yield strength, while increasing Mn in the Mn-5 alloy series increases the yield strength.

The mechanical properties for the alloys rolled at 850C and 950C are given in Figs. 33-39 at constant carbon concentrations. The alloys exhibit essentially the same characteristics as the 750C alloys. Elongation and γ stability increase as Mn percentage increases; yield strength first decreases with increased Mn content, as the percent α' present as-rolled decreases, and then increases, as the effects of solid solution hardening and carbon distribution become dominant. The increased γ stability to a α' strain-induced transformation with increased Mn content is illustrated in Fig. 40 for the Mn-5 850C test specimens. Here the α' transformation rate decreases for all alloys as the percent Mn is increased.

The effect of alloy content and transformation rate on uniform elongation is illustrated in the comparison of Mn-3 and Mn-5 alloys. The corresponding rates of formation for the 10-3 alloy versus the 12-3 alloy and for the 10-5 alloy versus the 12-5 alloy test specimens do not yield the same amount of uniform elongation (Fig. 41). The 10-3 tests have approximately the same transformation rates as the 12-3 tests, but only half the amount of uniform elongation. The 10-5 tests are found to exhibit greater rates of transformation than the respective 12-5 tests, even at approximately the same elongations. These discrepancies point out the complicated relationships between alloy content, γ matrix strength and stability, and α' strength. Small alloy changes can have profound effects on the stress-strain characteristics, as uniform elongation is a continuous battle between too little transformation, resulting in insufficient work hardening

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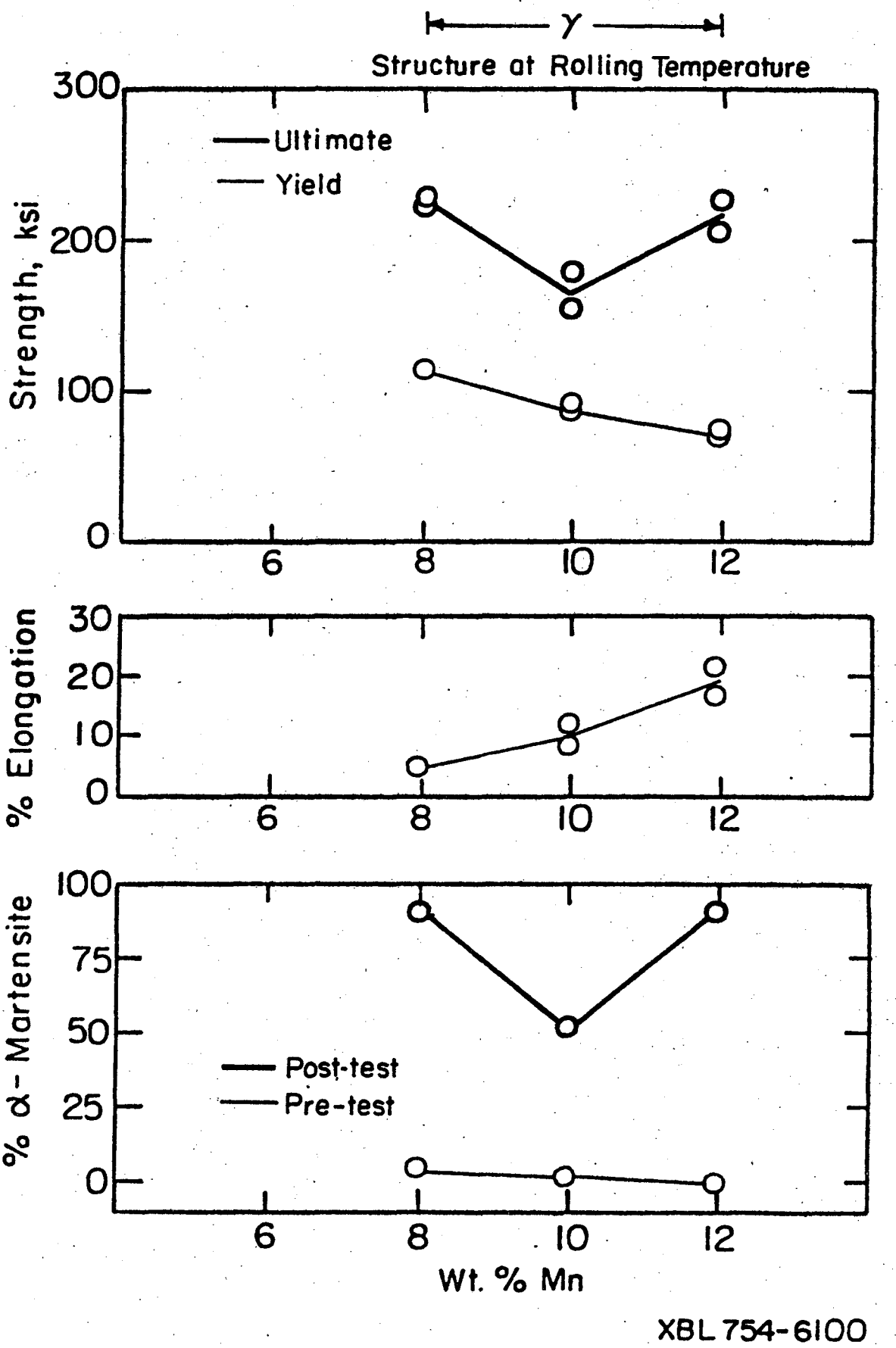

Fig. 33. Mechanical properties for Mn-3 alloys rolled at 850°C.

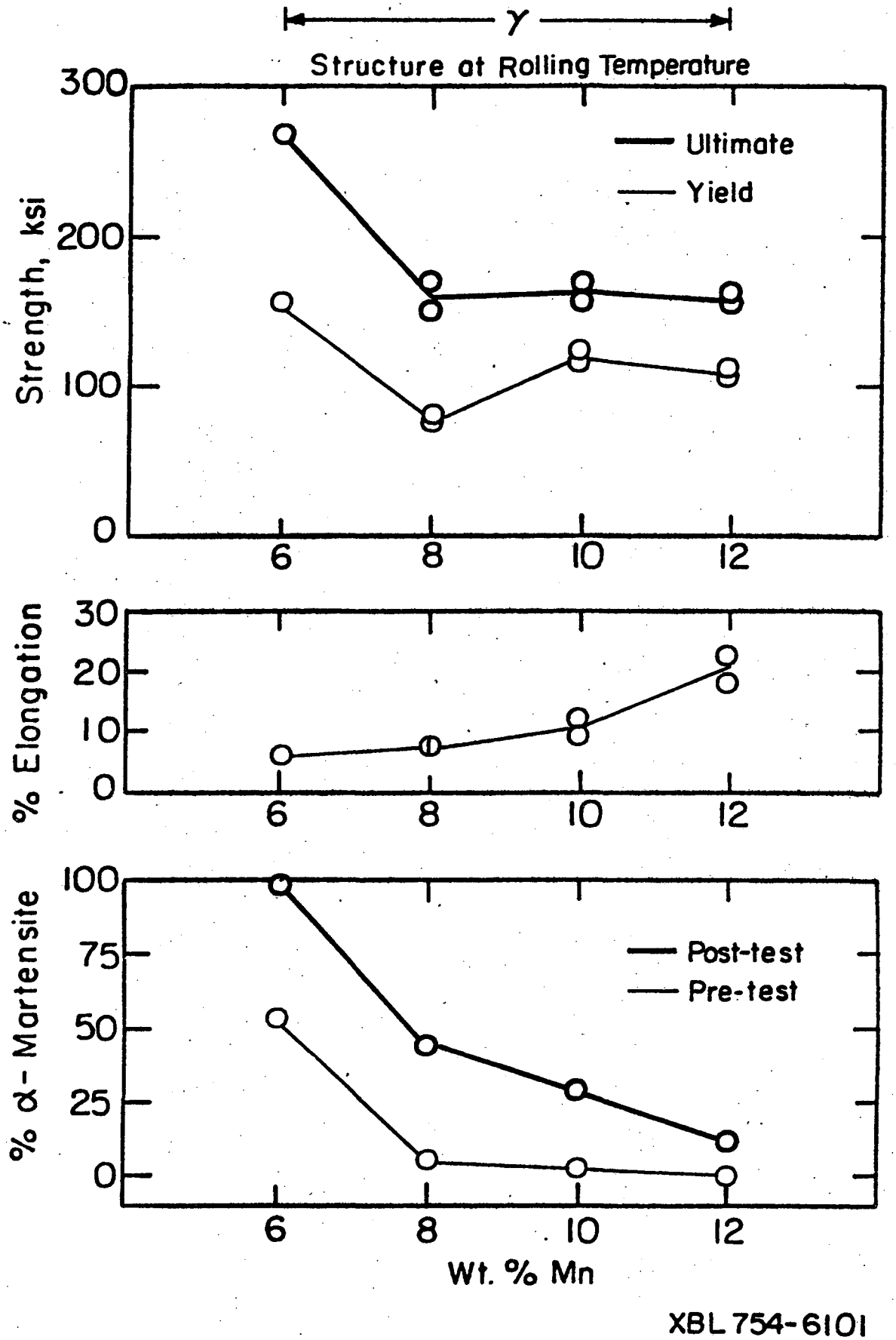

Fig. 34. Mechanical properties for Mn-5 alloys rolled at 850°C.

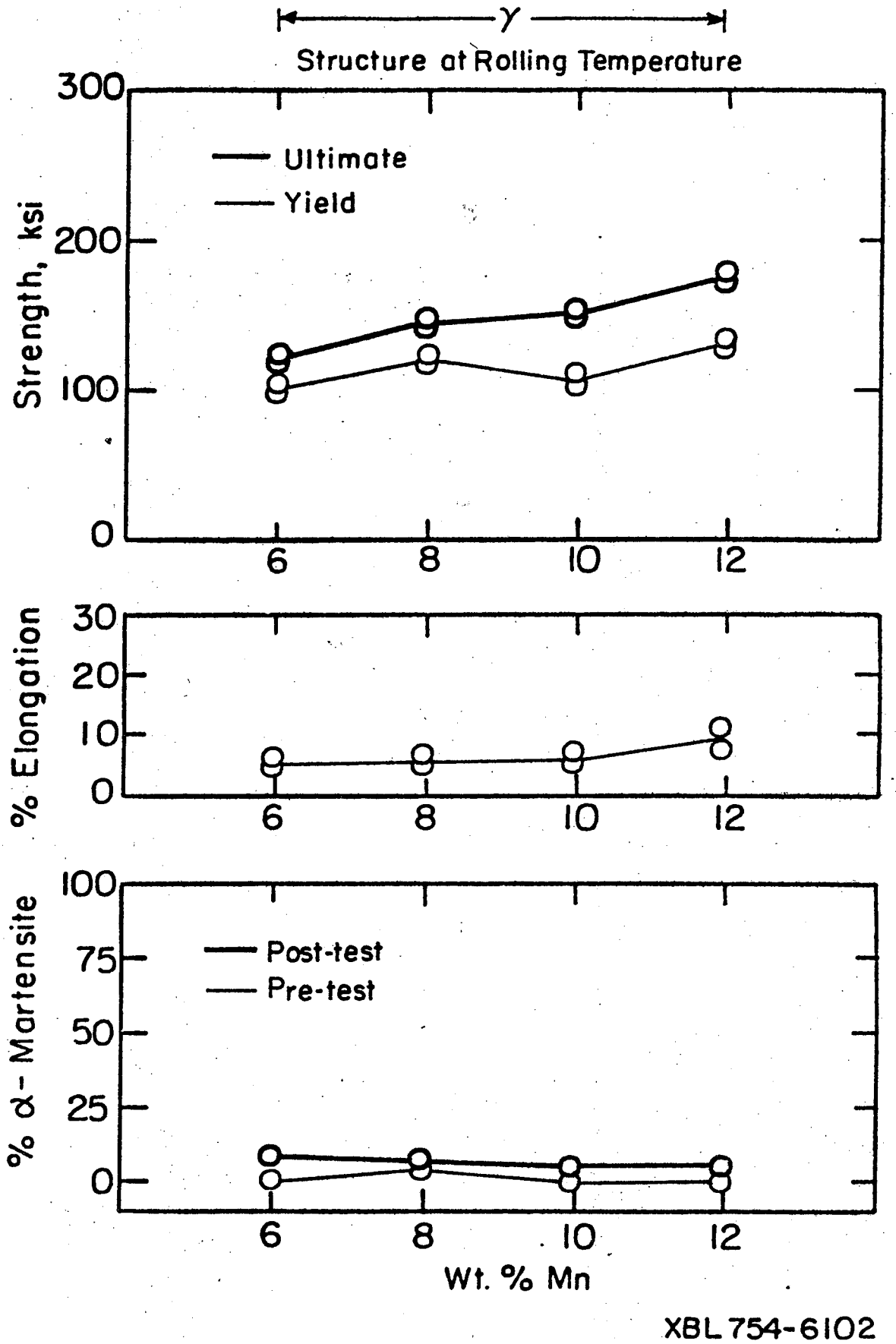

Fig. 35. Mechanical properties for Mn-8 alloys rolled at 850°C.

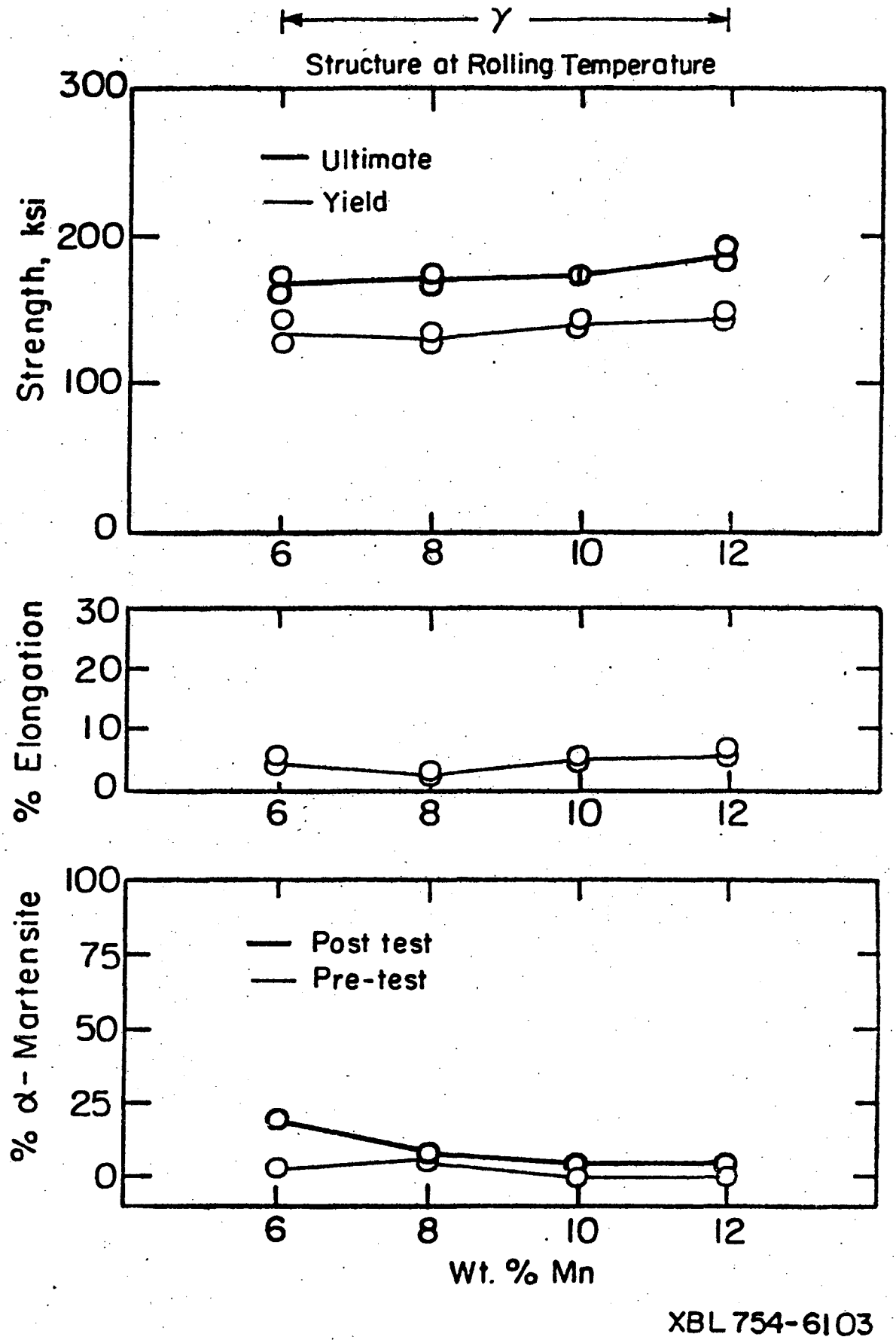

Fig. 36. Mechanical properties for Mn-10 alloys rolled at 850°C.

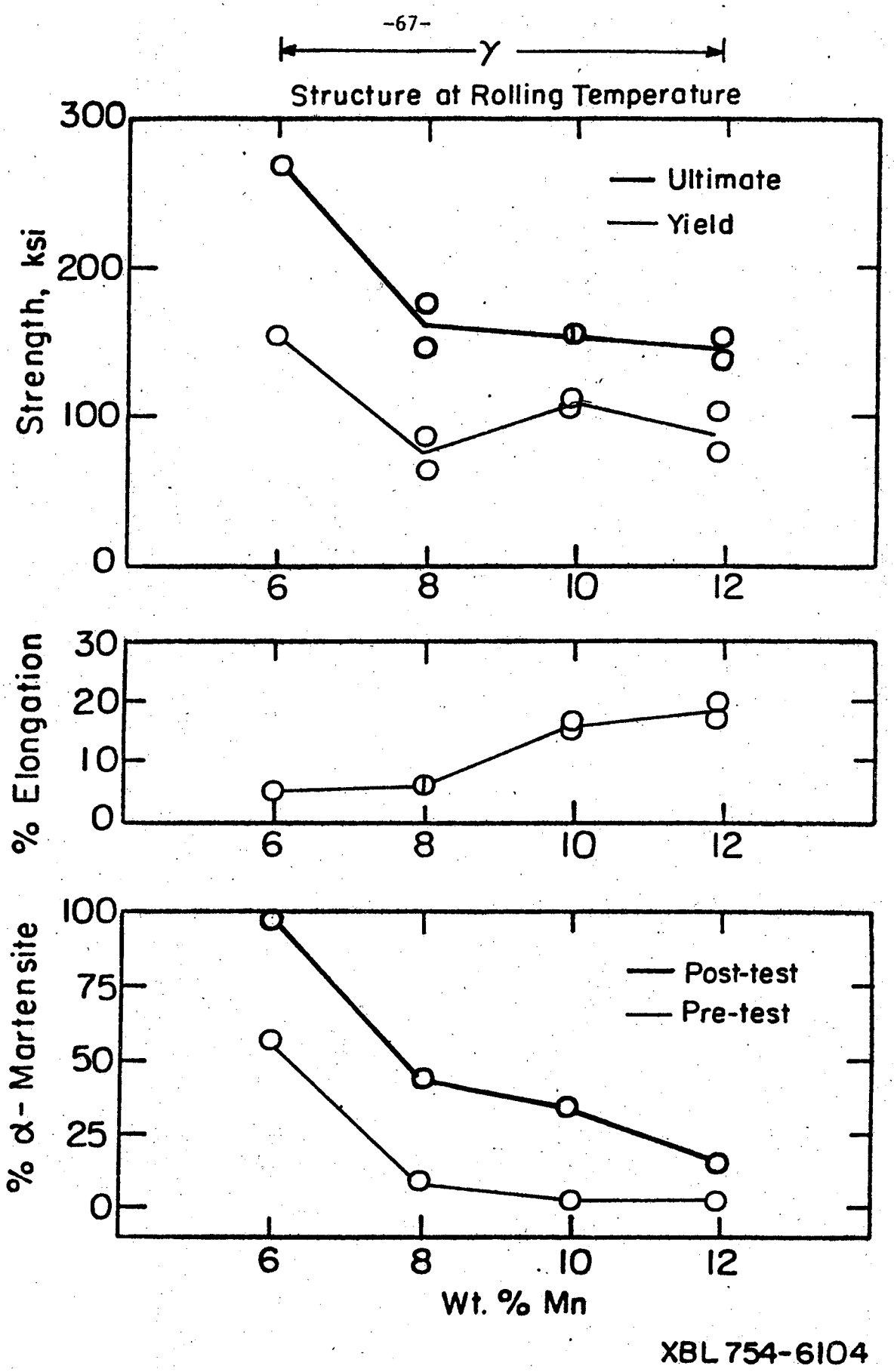

Fig. 37. Mechanical properties for Mn-5 alloys rolled at 950°C.

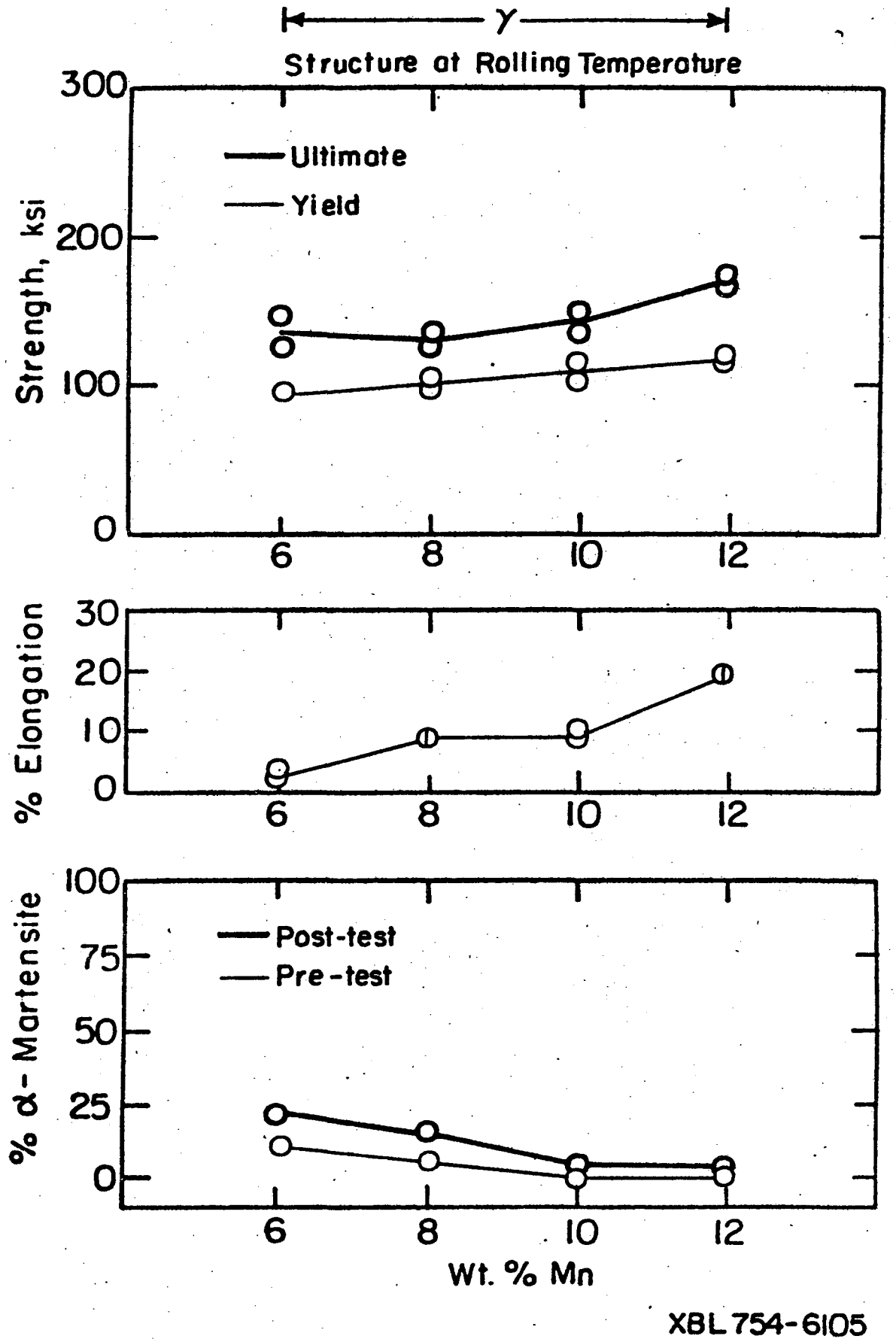

Fig. 38. Mechanical properties for Mn-8 alloys rolled at 950°C.

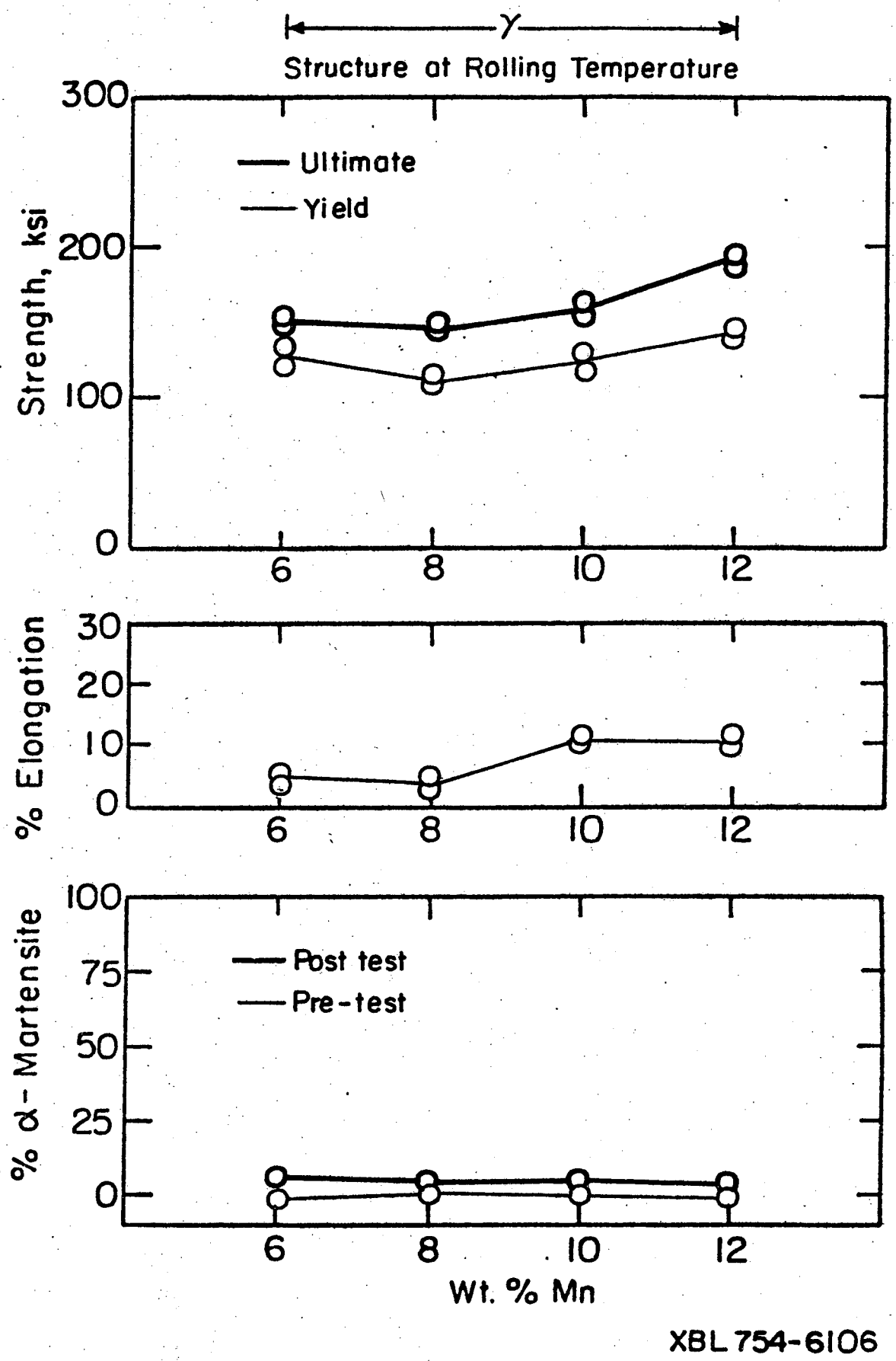

Fig. 39. Mechanical properties for Mn-10 alloys rolled at 950°C.

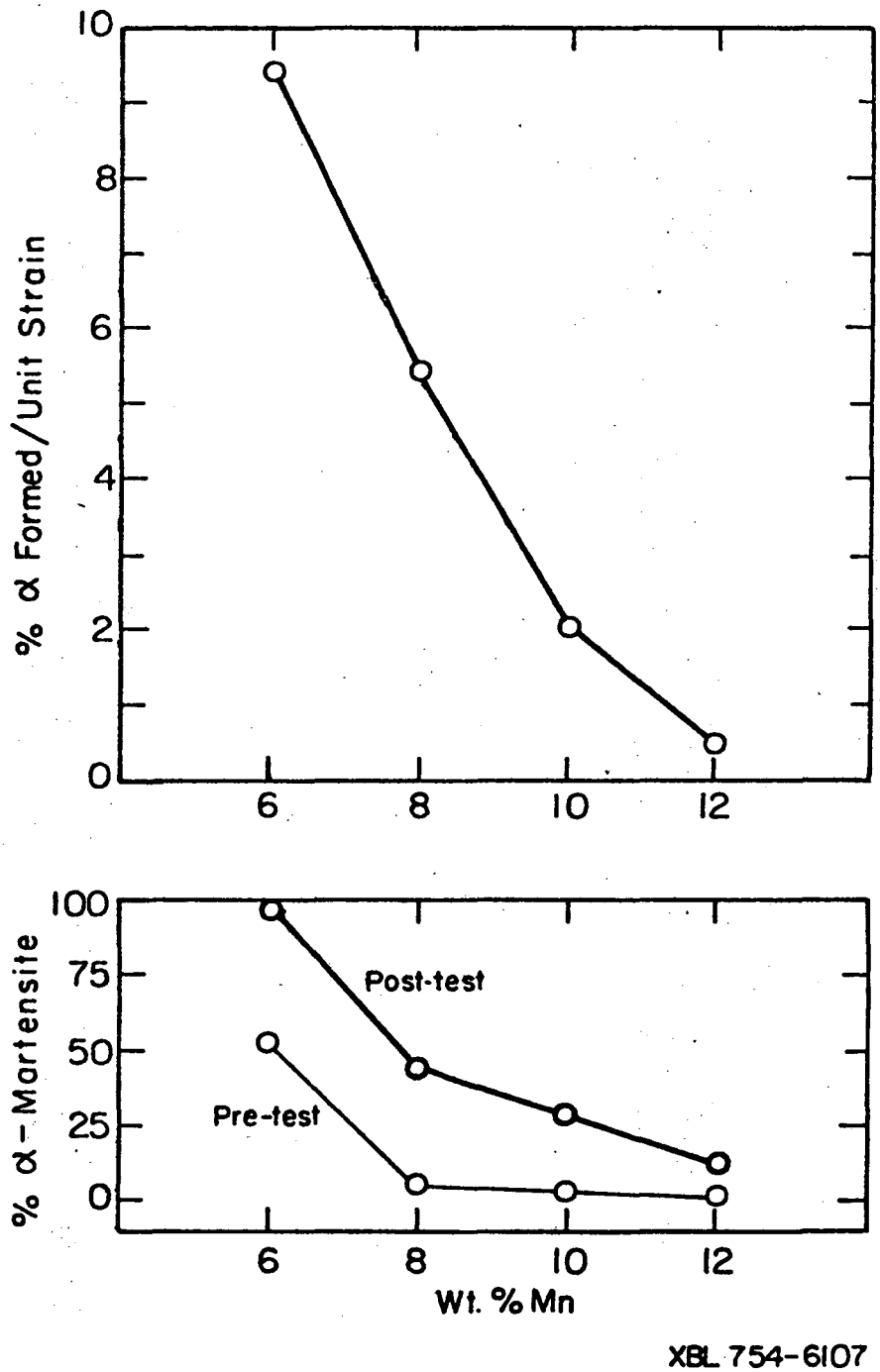

Fig. 40. γ to α transformation characteristics for Mn-5 alloy rolled at 850C.

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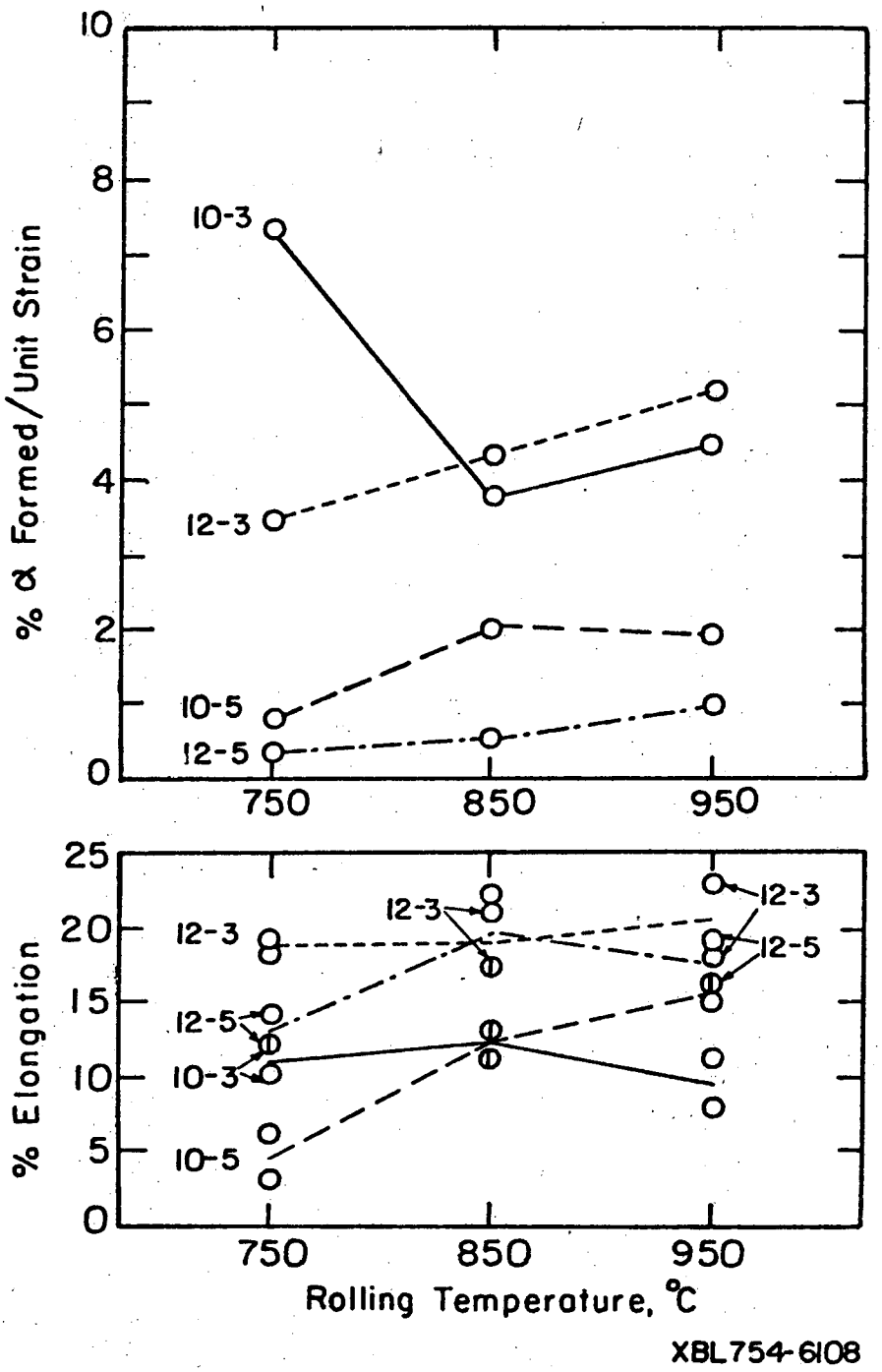

Fig. 41. γ to α transformation characteristics for Mn-3 and Mn-5 alloys.

to prevent necking, and too much transformation, resulting in localized embrittlement and subsequent specimen failure.

The large difference in carbon content between the 10-3 and 12-3 alloys (0.35 and 0.20% C, respectively) results in less of a decrease in γ stability in going from 12 to 10 percent Mn than would be expected from the 2-percent change in Mn. This increase in C percent in the 10-3 alloy also leads to an increase in α' strength, and a resultant greater amount of localized strain hardening. Thus, the same rate of formation in these alloys would be expected to result in a decrease in uniform elongation in going from the 12-3 to 10-3 alloy, due to too fast a work-hardening rate in the latter alloy.

The change due to the 2 percent Mn can be seen in the 12-5 and 10-5 alloys. Here the reduction in Mn reduces the stability of the γ with respect to an α' transformation and also decreases the strength of the γ matrix (Fig. 30). Thus, the decrease in Mn allows a higher rate of α' formation which is needed to yield the required rate of strain hardening, to result in high elongation, to compensate for the reduced matrix strain hardening component.[3]

6. Mechanical Properties Summary

The mechanical property results for the 850C and 950C specimens were than plotted as a function of changing carbon and changing Mn content (Figs. 42 and 43 respectively). Both the elongation and the yield strength values were plotted on the figures to allow easy determination of processing temperature and alloy composition for a given set of mechanical properties. The pre- and post-test percent α' are plotted in Figs. 44 and 45 for the 850C and 950C specimens using the same technique mentioned above. This allows one to include magnetic properties as an added alloying selection property.

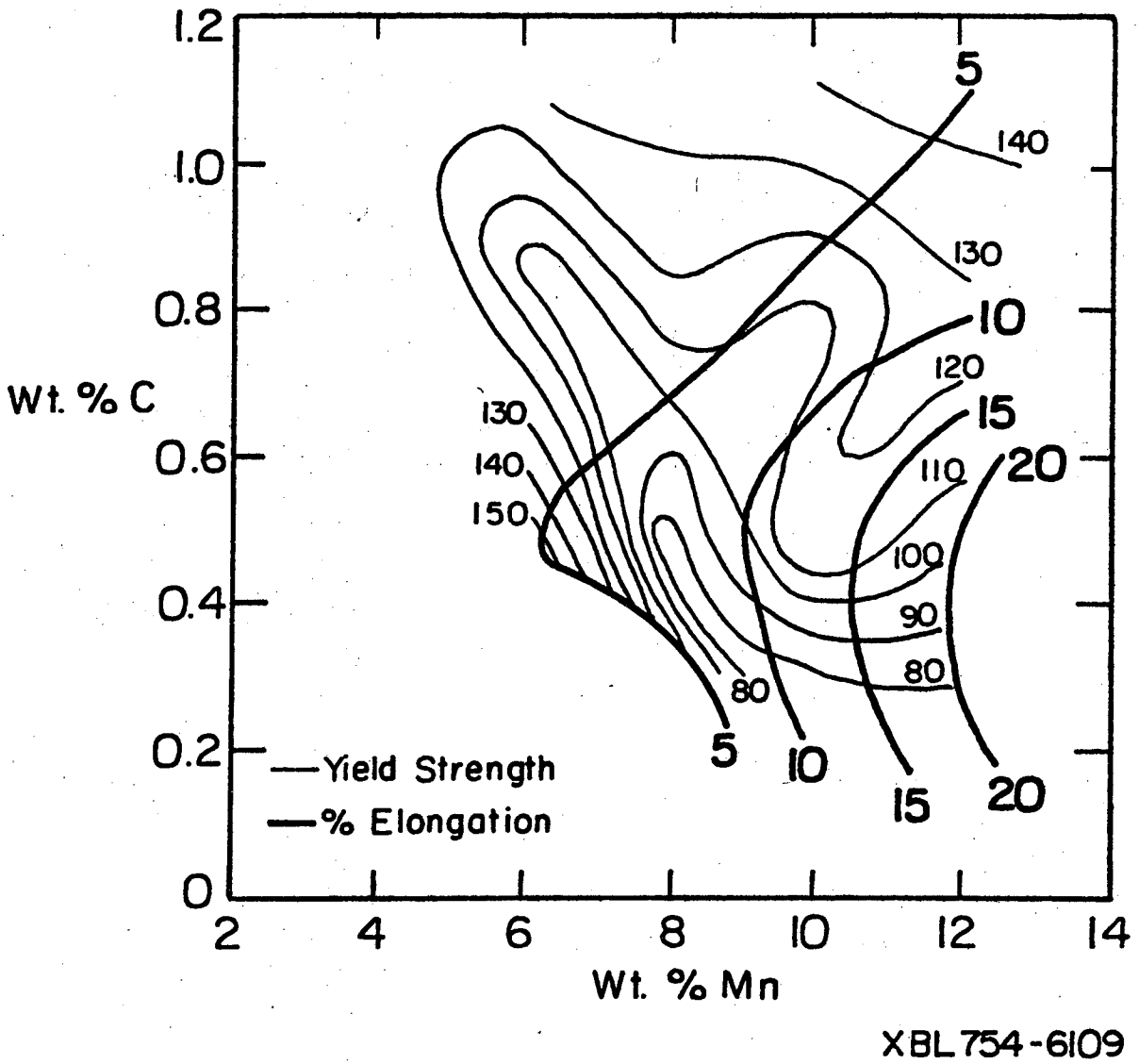

Fig. 42. Mechanical properties as a function of composition for alloys rolled at 850°C.

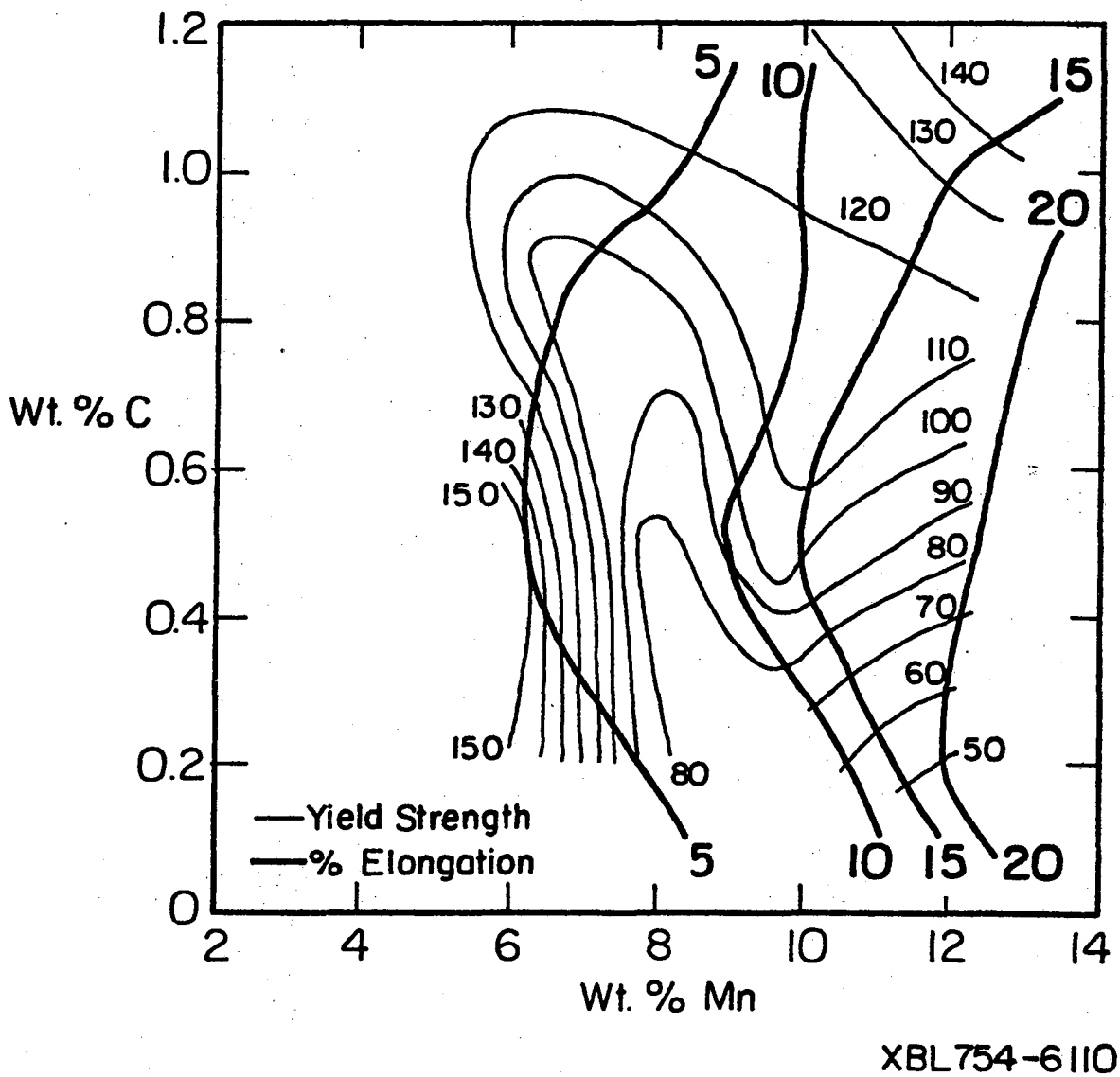

Fig. 43. Mechanical properties as a function of composition for alloys rolled at 950°C.

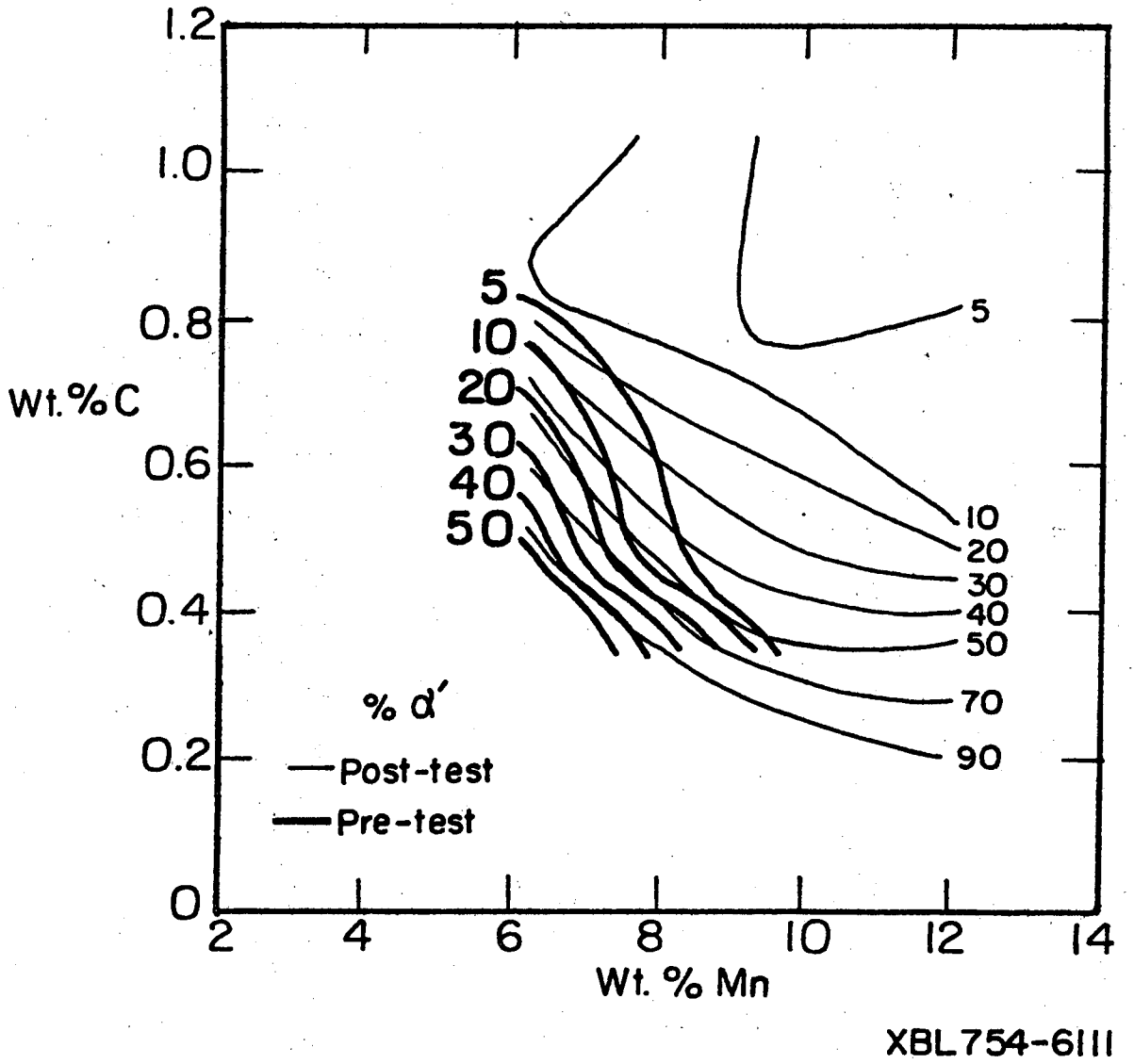

Fig. 44. Magnetic properties as a function of composition for alloys rolled at 850°C.

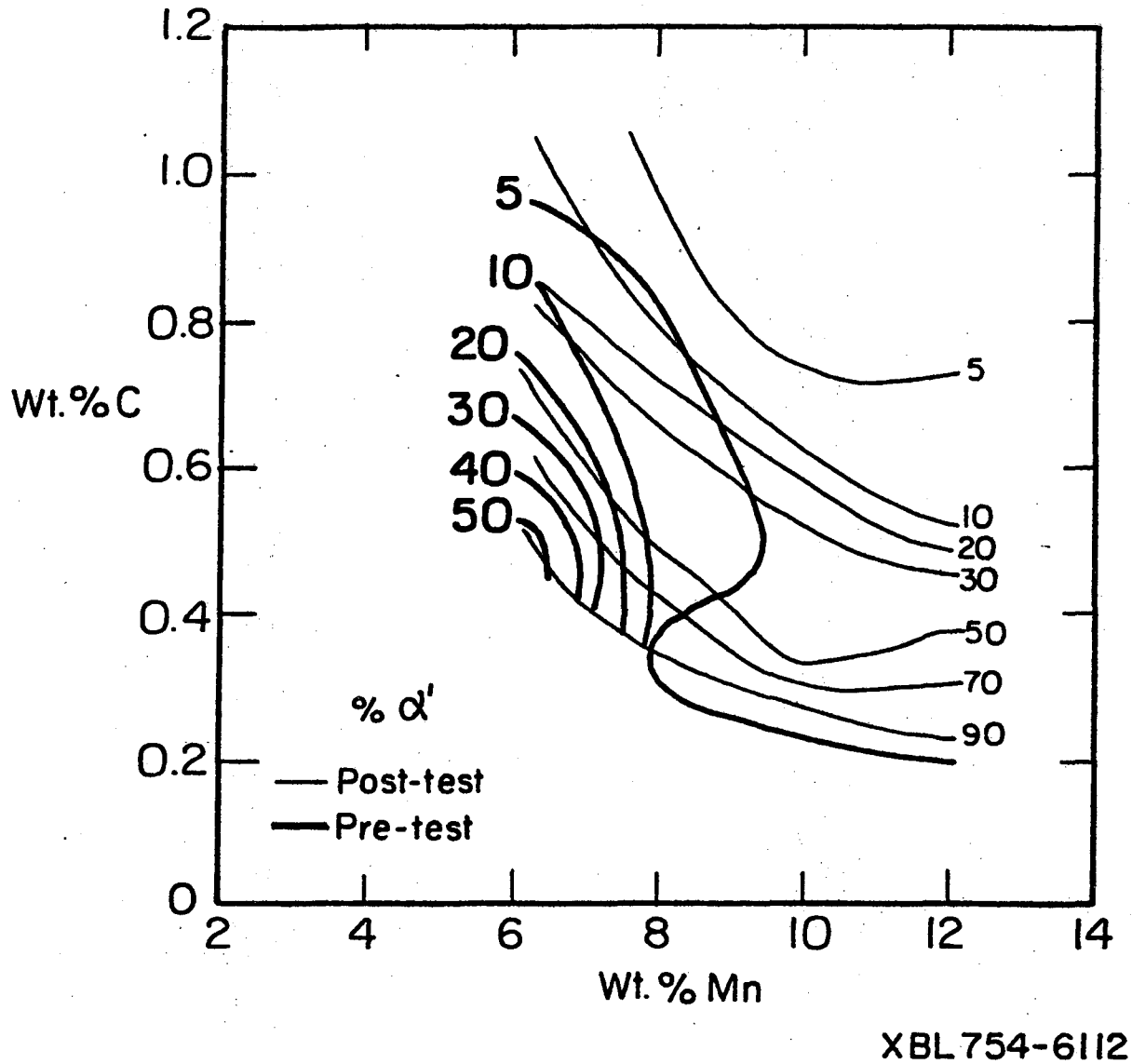

Fig. 45. Magnetic properties as a function of composition for alloys rolled at 950°C.

The combined analysis of these figures illustrates the general findings mentioned previously in the preceding sections. Both carbon and manganese stabilize γ; the yield strength increases as the percent as-rolled α' increases (decreasing carbon and/or manganese content) and also as the carbon content increases, with a minimum yield strength region where these opposing effects overlap. The elongation increases as the percent manganese or the rolling temperature increases; the yield strength decreases with increasing temperature.

B. Warm-Rolled Alloys

The alloys investigated in this phase of the study were three 12 wt% Mn alloys; they were the 12-3, 12-5, and 12-8 alloys. This series of alloys was used as it went from an alloy with a highly unstable γ matrix at room temperature, 12-3, to an essentially completely stable alloy, 12-8, based on the hot-rolling results presented in the preceding section.

The standard processing procedure for TRIP steel is to hot roll to the thickness required to yield the desired final thickness through an 80 percent reduction in thickness by warm rolling; the plate is then homogenized and quenched prior to warm rolling. It was found that none of the alloys selected for this study could be warm rolled at 450C or below while in the homogenized and quenched condition without exhibiting brittle fracture during rolling at thickness reduction values well below the desired 80 percent reduction; drastic reductions in uniform elongation after warm working has also been reported for other Fe-Mn-C alloys close to the compositions of those used in this work.[50] The 12-3 alloy exhibited the problem of transforming to α' during quenching and remaining martensitic

at the warm-rolling temperature, as opposed to the reversion to and stabilization of γ found during hot rolling. Yet the results from the last section indicate that these alloys hot roll without trouble and exhibit high elongations after rolling.

The processing procedure was therefore changed for these alloys to a procedure including a substantial thickness reduction by hot rolling prior to warm rolling, without a re-homogenization heat treatment as an intermediate step. The hot rolling was done at 950C, which is in the stable γ region for all alloys, and is also a rolling temperature at which all alloys exhibited high elongations. This was done to assure structure refinement without carbide precipitation for the high-carbon alloys. The percent hot-rolling reduction used was larger than that needed to yield a room-temperature metastable γ matrix for the 12-3 alloy. The specimens were then warm rolled at 400C.

The 950C hot rolling was done to refine the large grain structure present after homogenization. This processing tends to eliminate the heterogeneous, low-density potential precipitation-site structure and replace it with a high-density potential precipitation-site structure conducive to finely dispersed "homogeneous" precipitation.

The hot and warm rolling percent thickness reductions used for this portion of the study were 75 and 30, respectively; this resulted in a total thickness reduction of 83 percent. The results of this series of tests are shown in Fig. 46. It is evident that the yield strength rises almost linearly with increasing carbon content while both the elongation and percent α' formed during testing decrease. These results are compared to the properties found as-hot-rolled in Fig. 47. It can be seen that the

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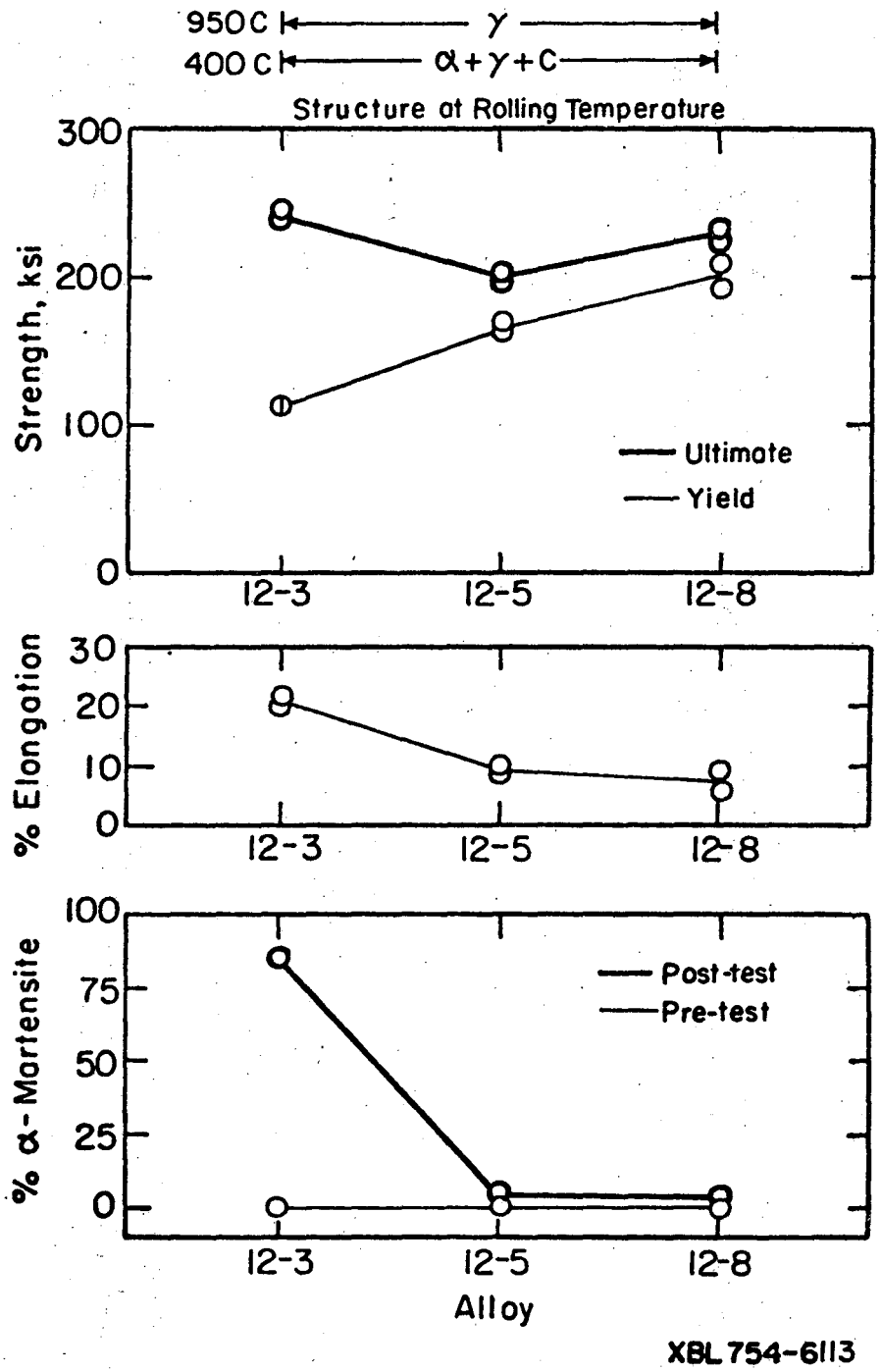

Fig. 46. Mechanical properties for 12-C alloys rolled at 950C+400C.

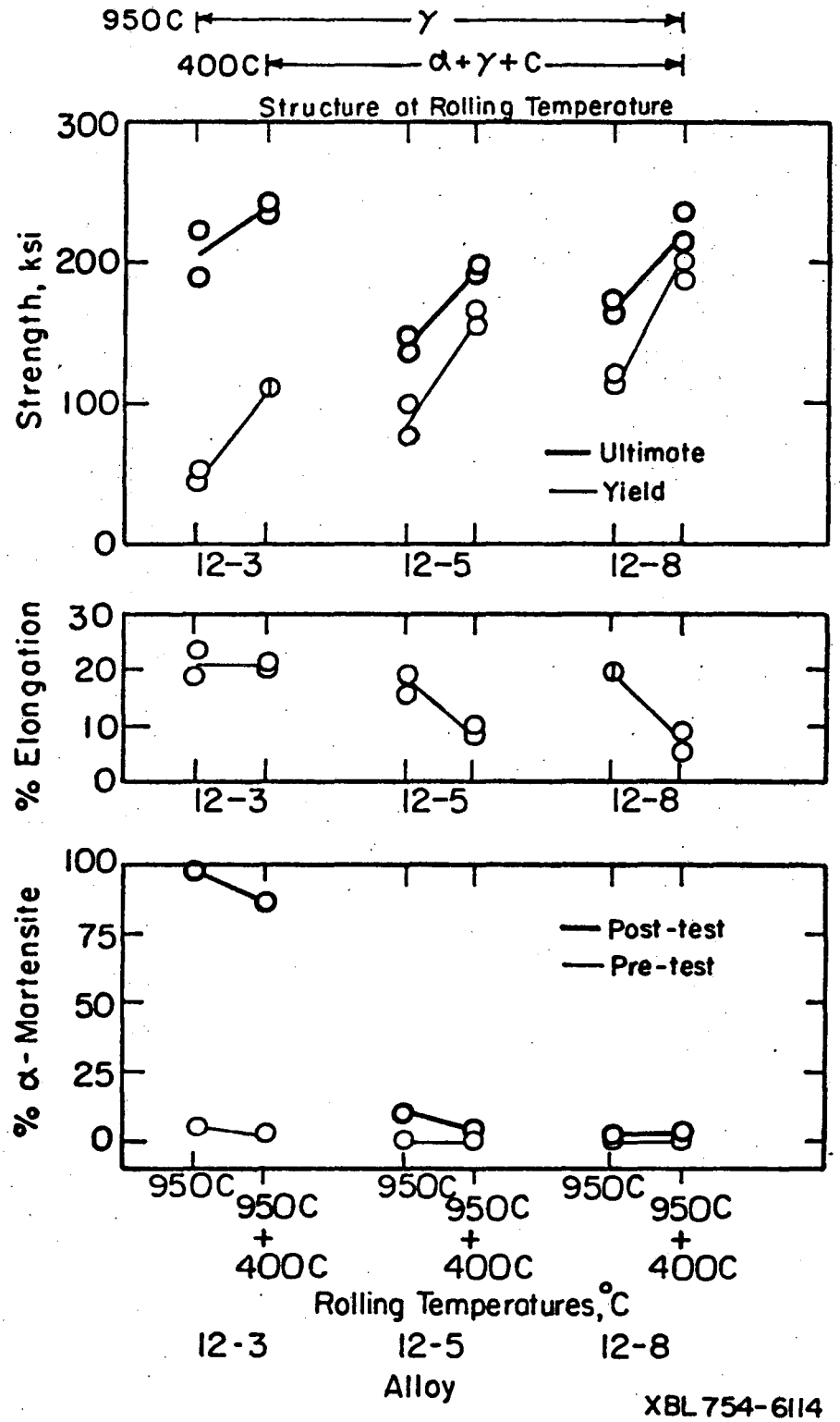

Fig. 47. Mechanical properties for 12-C alloys rolled at 950C and at 950C+400C.

hot-rolled elongations exhibited by all three alloys are high and essentially the same while the yield strengths again increase with carbon content and the percent transformation to α' decreases. Thus, the elongation values for the 12-3 alloy are essentially unaffected by the warm-rolling treatment even though the yield strength more than doubles; the elongation values for the other alloys are adversely affected by the warm-rolling treatment.

Three alloys exhibiting widely different as-hot-rolled stability were included in the test program to determine the effect of subsequent warm rolling on their γ stability. The hot-rolling results demonstrated that plastic deformation stabilized γ over transformation to α' when performed in the single phase γ region where no carbide precipitation was expected. The warm rolling was done at temperatures where precipitation was expected, and thus carbide depletion by precipitation was expected to decrease γ stability and compete with the plastic deformation stabilizing effect.

The percent α' transformed during straining, after warm rolling, was found to decrease for all alloys, as compared to the hot-rolled results (Fig. 47). However, as the amount of elongation for the 12-5 and 12-8 warm-rolled specimens also decreases, as compared to that for the respective hot-rolled alloy, it is hard to determine the relevance of their percent decrease. The 12-3 alloy did not show a decrease in elongation, and also exhibited a decrease in post-test percent of α' transformed after warm work. Thus, the composite results indicate that the stabilization of the γ matrix by plastic deformation is the predominant effect.

The 12-3 alloy was the only warm-rolled alloy exhibiting substantial transformation to α' or high uniform elongation. The constant change in alloy yield strength as a function of carbon content indicated that it did

not exhibit premature yielding. From these preliminary results the 12-3 alloy was determined as the alloy to study in order to produce a high-strength TRIP steel.

The 12-3 alloy was then hot rolled at 750C to determine the effect of a change in hot-rolling temperature prior to warm rolling at 400C. The results are compared to those for the 950C pre-roll, along with the as-hot-rolled results, in Fig. 48. It can be seen that the as-hot-rolled properties are approximately the same for both temperatures except for the slight amount of α' present as-hot-rolled in the 950C specimen. This is probably due to decarburization taking place during hot rolling, decreasing the γ stability near the specimen surface. But the 750C-400C tensile properties are clearly superior to the 950C-400C properties, because of the increase in yield strength with essentially no decrease in elongation. The 750C low-temperature hot roll was, therefore, used for the rest of the study.

The effect of varying the 12-2 warm-rolling temperature after hot rolling at 750C was studied next. The temperature was varied from 500C down to 200C, where the alloy was found to exhibit a magnetic transformation after 30 percent warm reduction, as determined by a hand magnet after the specimen was quenched to room temperature; rolling was stopped at this percent reduction for the 200C specimen. The results of the various treatments are shown in Fig. 49. The 200C warm-rolling results are not applicable towards the development of a TRIP steel because the as-processed specimens exhibited a magnetic transformation product, and it was not possible to determine the as-processed α' content because the specimen thickness was too great to fit into the premeameter due to

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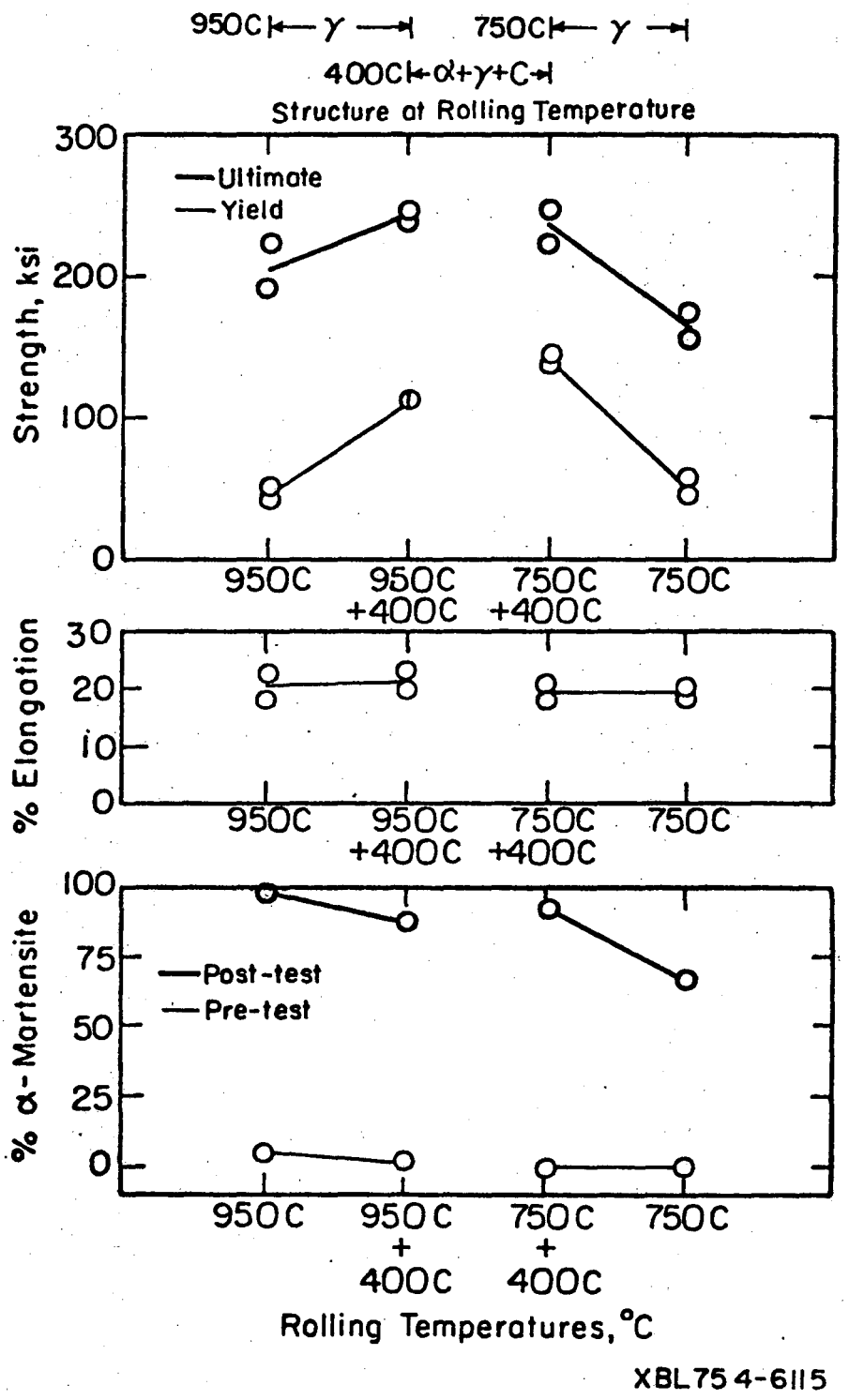

Fig. 48. Mechanical properties for 12-3 alloy rolled at 950C, at 950C+400C, at 750C, and at 750C+400C.

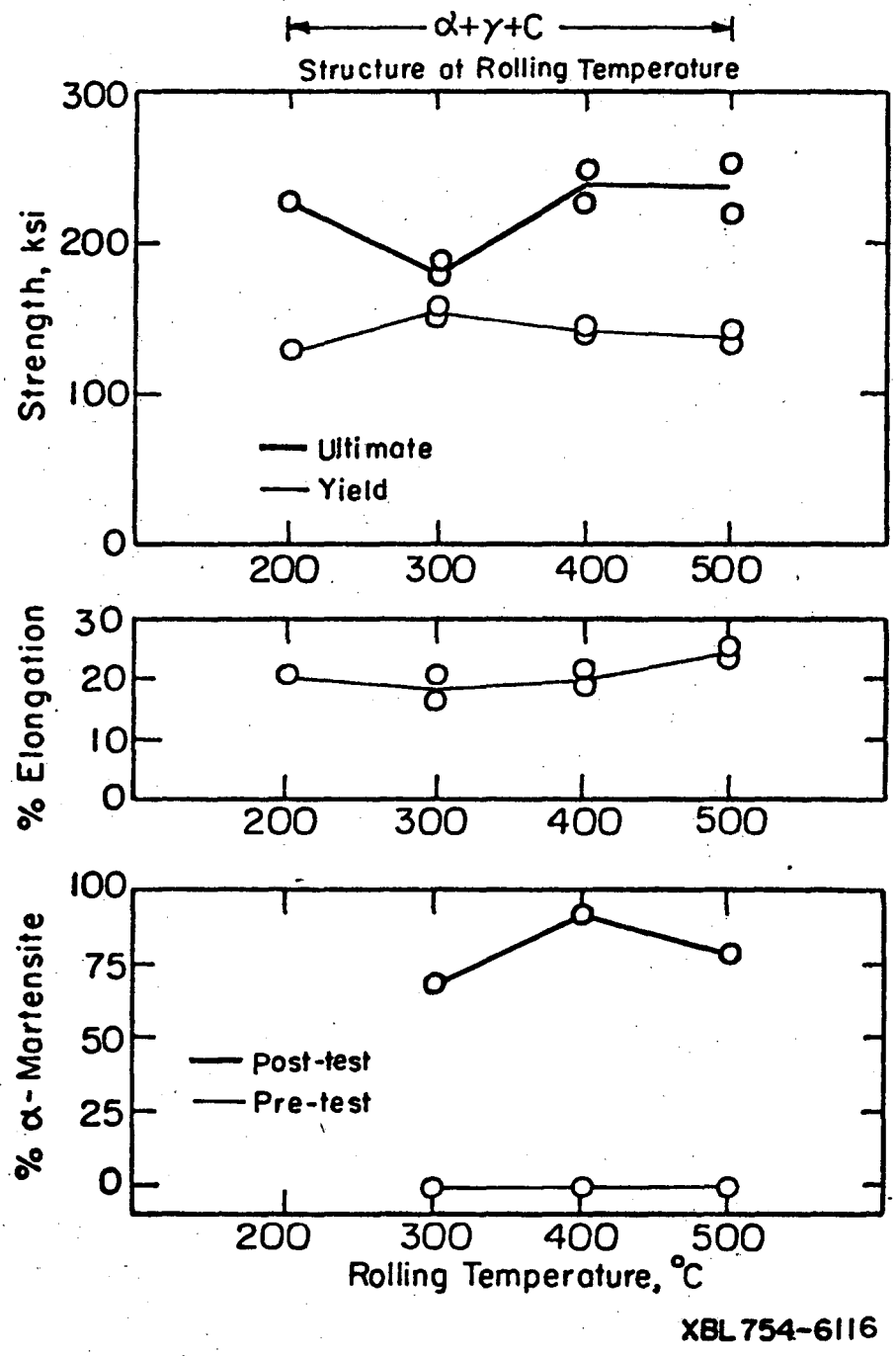

Fig. 49. Mechanical properties for 12-3 alloy rolled at 750C+X00C.

the termination of the rolling process once the magnetic transformation was detected.

The specimens tested corresponding to warm-rolling temperatures between 300 and 500C all exhibited high yield strengths and elongations. Therefore, the warm-rolling temperature is not particularly critical for this alloy-processing technique. The yield strength tends to increase and the elongations decrease as the warm-rolling temperature decreases. The tests were run on specimens that had had a relatively large amount of their 80-percent reduction in thickness performed by hot rolling, and were processed during warm rolling by an interrupted rolling procedure to allow determination of α' formation during processing. The specimens were homogenized prior to low-temperature hot rolling, but not hot forged or hot rolled at a high temperature before homogenization, which would have assured complete breakup of the as-cast structure prior to low-temperature hot rolling, and which would be done in industrial processing. The next series of tests was run to determine the change in properties through approximating an industrially practical rolling schedule. The next figure (Fig. 50) presents these results.

Three sets of results are plotted in Fig. 50. The first set consists of the properties of the 750C hot-rolled and the interrupted 400C warm-rolled specimens reported above, with an initial hot-rolling thickness of 0.25 in.; the second set was given essentially the same processing except without the interrupted warm-rolling sequence. The third set of specimens was homogenized at 1200C and quenched, hot upset forged at 1050C to 0.75 in. thick, re-homogenized at 1100C and then transferred to the hot-rolling-furnace at 750C (both homogenization treatments were done to all specimens

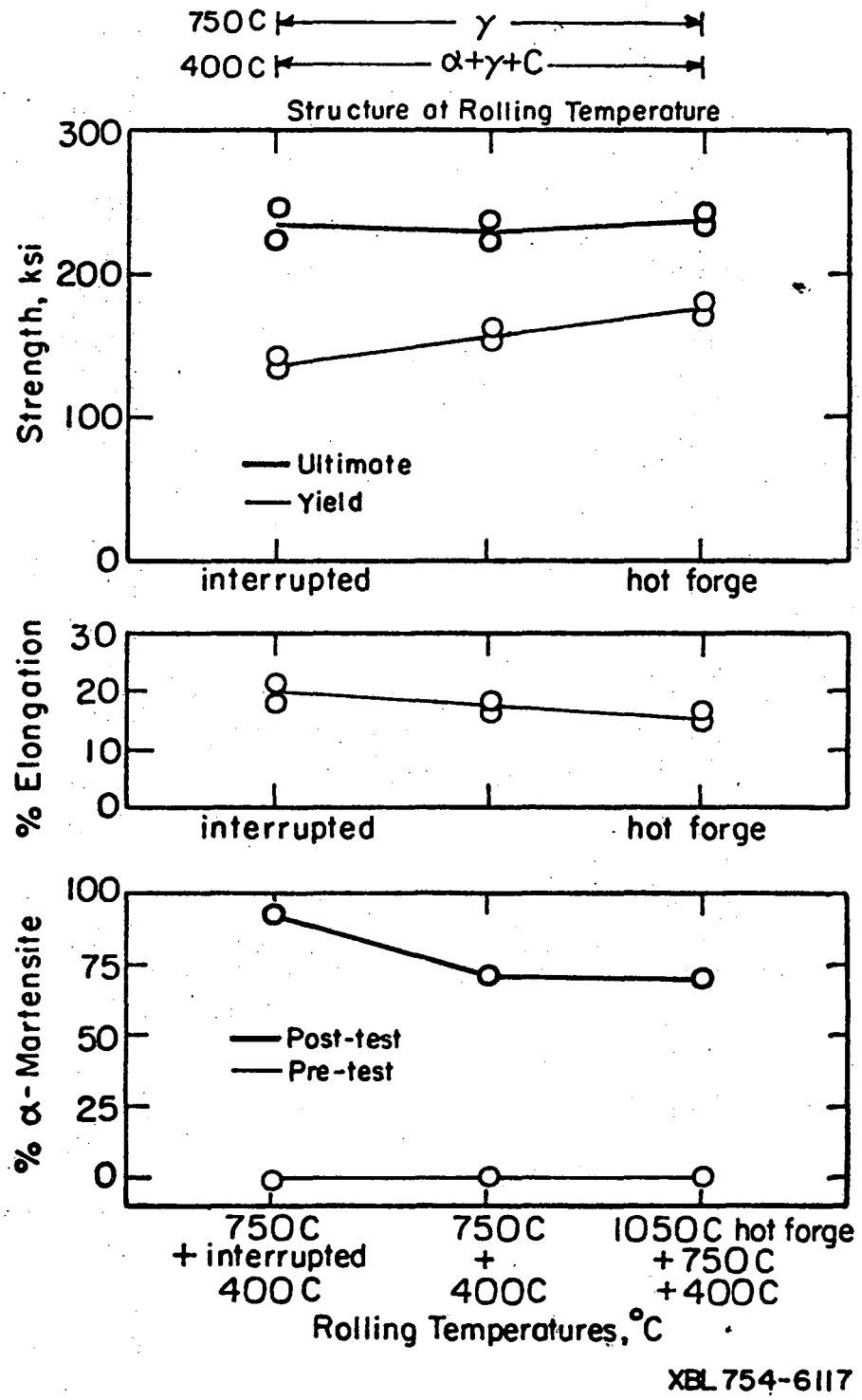

Fig. 50. Mechanical properties for 12-3 alloy rolled at 750C+400C with various processing procedures.

and were performed with the specimens doublebagged in stainless steel foil, as explained in the experimental section). The specimens were then hot rolled 66 percent to 0.25", quenched, and further warm reduced 80 percent in thickness at 400C. The resulting yield strengths are found to increase considerably as the processing approaches that of a potential industrial process with only a slight decrease in elongation. The last processing procedure results in a TRIP steel capable of exhibiting a yield strength of 180ksi and an elongation of 20 percent.

A further study was carried out to determine the effect of replacing warm-rolling reduction with hot-rolling reduction. The results of this are presented in Fig. 51. The total reduction in thickness, based upon the pre-hot-rolled thickness, was kept constant at 94 percent; the amounts of hot-and warm-rolling reductions were varied as shown on the figure. It can be seen that the amount of reduction achieved by warm rolling can be substantially reduced for a slight decrease in yield strength; this type of processing optimization would greatly reduce the warm-roll wear, and subsequently reduce production costs.

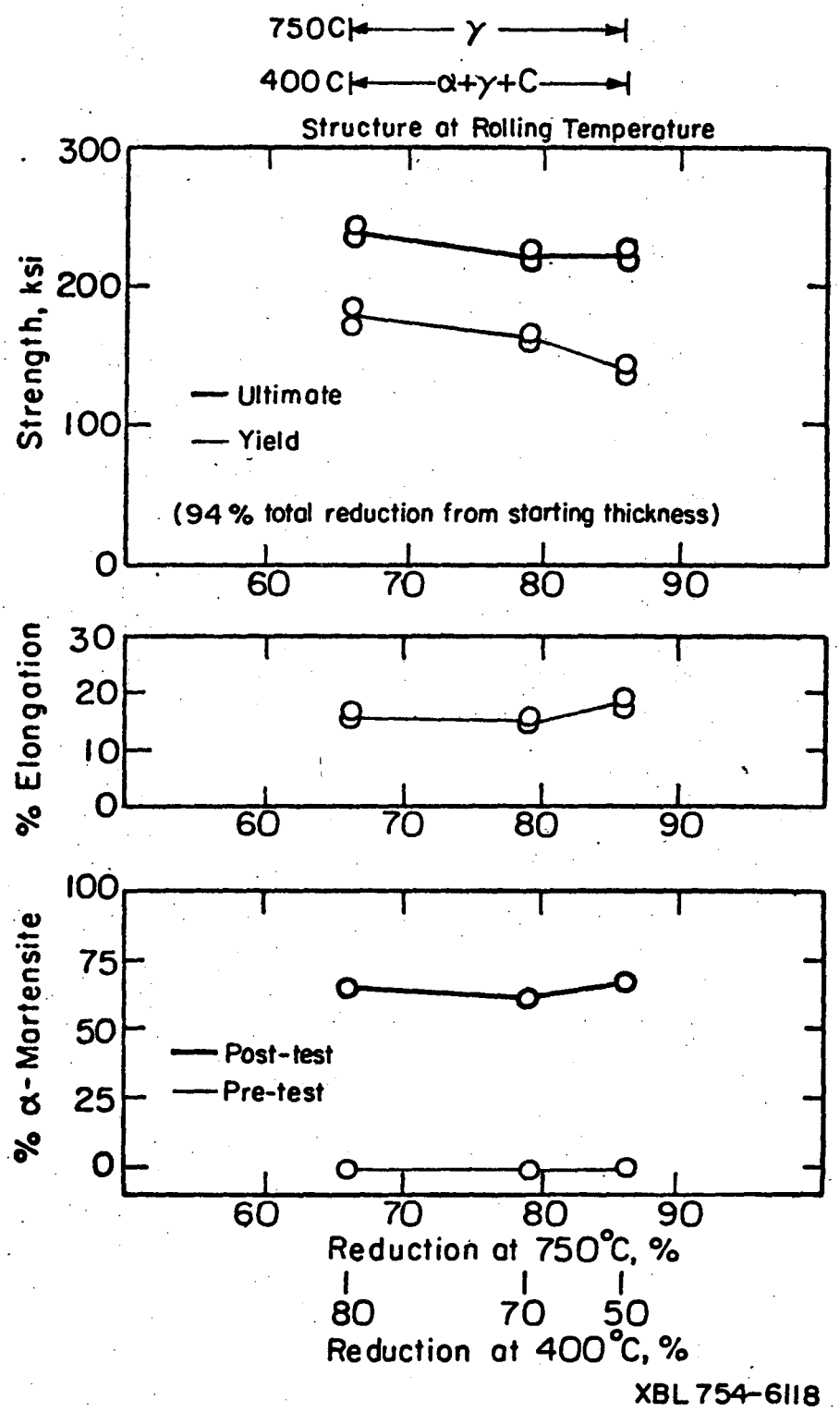

Fig. 51. Mechanical properties for 12-3 alloy rolled X% at 750C+Y% at 400C.

IV. DISCUSSION OF RESULTS

This section, as is the Experimental Results section, is broken down into two sub-sections by processing procedure. The first sub-section contains the discussion of the hot-rolling results; the second contains the discussion of the warm-rolling results.

A. Hot Rolled Alloys

The results of this section show that a low-cost, medium-strength TRIP steel of 100ksi+ yield strength can readily be produced. The processing involves only low-temperature hot rolling (750C to 950C) and thus eliminates the expensive warm-rolling process needed for high-strength TRIP steels. This alloy development is specifically aimed at the majority of industrial applications of steel which utilize low- and medium-strength steel, as opposed to the specialized-use area covered by standard high-strength TRIP steels. The vast majority of steel tonnage used by industry is at yield-strength levels below 150ksi. In general industrial use, a steel exhibiting a yield strength of 100ksi or greater would be called a high-strength steel; a steel over 150ksi yield would be called an ultra-high-strength steel, and would be used only for very specialized applications.

The mechanical properties of these alloys are superior to present hot-rolled alloys now used; typical hot-rolled alloys exhibit yield strengths of 60ksi, approximately half that exhibited by some of these experimental alloys, with the same elongations; these properties are also considerably above those found in work aimed toward developing modified hot rolling procedures for this alloy system.[46,55-57] The detailed study which varied both composition and rolling temperature established that these alloys are

relatively insensitive to slight variation of either parameter, thus showing that consistent properties in industrial heats could be achieved even with the composition and processing variations inherent in industrial fabrication procedures.

The γ matrix of these alloys increases their desirability over conventional medium-strength α steels. It is expected that this type of alloy could be used to solve the embrittlement problems, found in the petroleum and chemical-processing industry, that limit most steels used in these applications to less than 100ksi yield strengths. The high-strength TRIP steels have been found to be considerably more resistant to embrittlement than standard α-matrix steels with comparable yield strengths, and require extremely severe conditions to induce embrittlement.[50-53] Because susceptibility to embrittlement is known to decrease as strength level decreases, these medium-strength alloys are predicted to be even more resistant to embrittlement, and considerably more resistant than standard steels of comparable strength level.

The experimental results reveal that some of the hot-rolled medium-strength steels tested remain essentially completely austenitic while exhibiting high elongation. The amount of transformation to α' during testing was found to vary considerably with alloy content; the low-carbon-content alloys exhibited almost complete transformation, while the high-carbon content alloys exhibited almost completely stable γ. This change in transformation products takes place with almost no change in elongation and an increase in yield strength, if the correct processing parameters are used.

Environmental embrittlement of high-strength TRIP steel has been found to take place only once a finite amount of α' has been formed by plastic flow, and the same is expected of these medium-strength TRIP steels. Thus, this group of stable γ Fe-Mn-C alloys holds out the promise of complete immunity to environmental embrittlement, although the corrosion resistance and longtime carbide stability would have to be investigated.

B. Warm-Rolled Alloys

The results of this section show that a low-cost, high-strength TRIP steel can be readily produced utilizing the Fe-Mn-C ternary alloy system; the alloy composition used in this study was Fe-12 wt% Mn-0.2 wt% C. This alloy was able to attain a γ matrix yield strength of 180ksi even though no strong carbide-forming alloy elements were present; the carbon content of this alloy was also lower than that usually used in TRIP steel alloys (0.3 wt% C). The 180ksi yield strength was accompanied by 20 percent uniform elongation.

The processing recommended for industrial production of the alloy would be as follows:

1. Homogenize billet.
2. Hot-forge and/or high-temperature hot roll billet.
3. Homogenize billet.
4. Cool billet to temperature for low-temperature hot rolling and hot roll (750C and 70% or more reduction is recommended).
5. Quench to warm-rolling temperature or below.
6. Warm roll to final thickness (80% reduction in thickness recommended).

The inclusion of a low-temperature hot rolling step directly after high-temperature hot rolling and subsequent homogenization (done without an

intermediate cooling to room temperature between homogenization and low-temperature hot rolling) is a major change from the standard TRIP steel processing procedure.

This procedure was found to be necessary for this alloy system due to the need for a given amount of low-temperature hot rolling to yield a room-temperature γ matrix for the 12-3 alloy. But the author believes that this type of processing could well improve the properties of a standard TRIP steel alloy as well, and may present a way to overcome the current upper TRIP steel yield-strength ceiling.

TRIP steels exhibiting substantial uniform elongation are currently limited to yield-strength levels below 275 to 300ksi.[64,65] The results of this section indicate two different ways to overcome this problem. The increase in TRIP steel yield strength has so far been achieved through an increased carbon content, but the results of this section show that a substantial increase in yield strength can be achieved simply through the inclusion of a low-temperature hot-rolling treatment prior to warm rolling. Thus it may be possible to increase the yield strengths of present TRIP steels by an intermediate low-temperature hot-rolling treatment.

The pre-warm-rolling treatment is also found to increase the deformability during warm rolling of high-carbon alloys. This is explained as due to the formation of a refined substructure -- which would be expected to exhibit uniformly fine homogeneously distributed potential precipitation sites -- while the steel is still in the γ stability region. This eliminates the possibility of formation of a network of precipitates, which may form around the coarse grains found after a homogenization treatment, during the cooldown to room temperature, heatup to warm-rolling temperature, and the first few reduction-in-thickness rolling passes. The author postulates that

this same type of problem may be affecting the high-carbon standard TRIP steel alloys. An initial heterogeneous precipitation network may be achieved before warm-rolling substructure refinement takes place, and the network may not be completely broken up by subsequent warm rolling. Thus a low-temperature hot rolling treatment immediately after homogenization, and prior to cooldown and warm rolling, would be expected to result in a high-carbon TRIP steel with a yield strength over 300ksi and increased elongation.

The inclusion of a low-temperature hot-rolling step in the standard TRIP steel processing also opens up the possibility of decreasing production costs, as mentioned in the presentation of the results for the last figure. The high production costs of TRIP steel come mainly from the requirement for large-capacity rolls for warm rolling, and their subsequent wear. The feasibility of substituting hot reduction for warm reduction has been shown, and becomes even more attractive if the prediction of increased yield strength due to the inclusion of a low-temperature hot-rolling step is true. By replacing present warm-rolling reduction by as much hot-rolling reduction as is possible for a given yield strength, costs for TRIP steels now in production can be reduced.

This study was confined to the Fe-Mn-C ternary system to allow the property changes at room temperature to be correlated with changes in the phase-stability regions where the respective processing was carried out. The low alloy content of the 12-3 alloy indicates that the effect on phase stability from the addition of other alloying elements for special purposes -- such as increasing the corrosion resistance by adding Cr, or the yield strength by adding a strong carbide-forming element -- could be

compensated for by an increase (or decrease) in manganese and/or carbon content while still retaining a basically iron-based alloy. Thus, the results of this study indicate that a new family of low-cost, high-strength TRIP steel alloys, with tailor-made properties, based on the Fe-Mn-C alloy system is within the grasp of industry.

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V. CONCLUSIONS

The results reported in this thesis show that the Fe-Mn-C system can be utilized for the production of low-cost TRIP steels which vary in yield strength from 75 to 180ksi and exhibit 20 percent uniform elongation. These steel alloys contain 10-12 wt percent Mn and 0.2-1.0 wt percent C, and achieve their high γ-matrix yield strengths without utilizing strong carbide formers. The use of low-cost Mn and the lack of high-cost strong carbide-forming alloying elements, and the lack of high-cost Ni as a principal alloying element (both currently used in TRIP steel), results in a substantial reduction in raw material costs for this proposed TRIP steel alloy system.

The production costs of both these TRIP steels and the present alloys can be reduced below those now encountered, through the inclusion of a low-temperature hot-rolling step prior to warm rolling. The hot-rolling temperature used would be just into the given alloy's stable γ region, and would be done directly after an austenitizing treatment without an intermediate cooldown below the hot-rolling temperature. This extra processing treatment is expected to increase the resultant room-temperature yield strength for a given alloy without loss of elongation, and thus decrease the production cost for a steel of given yield strength by decreasing the amount of warm rolling needed to achieve the given yield strength; warm rolling is a high-cost operation in comparison to hot rolling as it requires higher-capacity rolls and results in accelerated roll wear. The inclusion of the hot-rolling step into the standard TRIP steel processing

procedure is considered a major change in production processing that will result in immediate processing benefits.

The hot rolling prior to warm rolling is also predicted to present a way of increasing the yield strength of TRIP steels over the practical limit of 275-300ksi now in existance; TRIP steels with yield strengths over 300ksi have been produced but exhibit drastic reductions in uniform elongation. As mentioned above, the inclusion of this additional processing step is expected to increase the yield strength without loss in elongation of the TRIP steels now in production, if the final warm-rolling reduction is kept constant. Thus, the inclusion of this additional processing step may in itself accomplish the desired results.

The hot-rolling step was also found to increase the ease of deformation of high-carbon-content alloys. Thus, it is predicted that inclusion of a hot-rolling step prior to warm rolling will increase the elongations exhibited by the 300ksi+ yield strength alloys which were mentioned above, as these steels utilized increased carbon to increase their yield strength. This assumes that the problem is mainly a carbide-precipitation problem and not a γ-stability problem; this is indicated by the results of this thesis.

It was found that rolling just into the γ region increased the amount of room-temperature metastable γ for low-carbon-content alloys and increased the deformability of high-carbon alloys. Hot-rolled alloys exhibiting elongations over 15% required a minimum of 10 wt% Mn but varied in carbon content between 0.2 and 1 wt % carbon, depending on the composition and processing. The composition of the high-strength TRIP steel

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developed in this study was Fe-12 wt % Mn-0.2 wt % C, but the author predicts a whole family of high-strength TRIP steels could be based on this composition, including alloys with higher carbon content (with possibly the addition of Cr or Si or a decrease in Mn to offset the change in γ stability), to increase the yield strength above 180ksi. The possibility of the development of a medium-strength austenitic steel exhibiting almost complete stability with respect to a strain-induced phase transformation using the Fe-Mn-C system was also demonstrated.

The high-elongation hot-rolled steels break down into two alloy families. The low-carbon-content alloys exhibit the expected TRIP steel strain-induced transformation to α'; the high-carbon-content alloys exhibit decreasing amounts of transformation until an essentially stable γ matrix is reached. Thus, this research indicates that resistance to environmental embrittlement can be improved through the development of low-cost medium-strength TRIP steel, and possibly eliminated by low-cost, completely austenitic medium-strength steels based on the Fe-Mn-C system.

The need for industrial alloys to withstand increasingly severe environments and the economical desirability of increased yield strengths results in the need for the development of new low-cost embrittlement-resistant alloys. The results of this thesis indicate that, through the correct thermal-mechanical processing, the Fe-Mn-C alloy system can be used as a basis for a new series of low-cost TRIP steel and austenitic alloys covering the entire strength spectrum currently required by industry. Although this is one of the first steel alloy systems to be investigated, increased understanding of the factors determining phase stability continue to make it one of the most promising.

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