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QUANTITATIVE INVESTIGATION OF STRAIN INDUCED STRENGTHENING IN STEEL

Bertrand de Miramon
(M. S. Thesis)

September 1967

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QUANTITATIVE INVESTIGATION OF STRAIN INDUCED STRENGTHENING IN STEEL

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ABSTRACT

The effect of the strain induced austenite → martensite transformation on the mechanical properties and especially on the ductility of some metastable austenitic steels has been investigated. The steels investigated were a new class of high nitrogen steels having yield strengths above 200,000 psi with 20 to 40% elongation. The results emphasize the important role of the austenite → martensite transformation in improving the ductility by preventing early necking, as has been mentioned by other authors.

The austenite → martensite transformation was detected by a magnetic method, assuming that the saturation induction (B_s) was proportional to the amount of magnetic phase (martensite).

The proportion of martensite in each specimen was measured during pulling using a permeameter attached to an Instron testing machine. The tests showed that martensite forms immediately after the yield point in a region of the specimen which is plastically deformed.

Some static measurements were also made on small portions cut from the tensile specimen. These measurements showed that the austenite → martensite transformation seems to occur only in the plastic region which traverses the specimen.

Finally a few continuous (but non-quantitative) tests were run and these showed that the serrations observed in some tensile curves can be correlated with discontinuities in the martensite transformation.

I. INTRODUCTION

Recently a new series of metastable austenitic steel combining high yield strength with good ductility has been investigated. Such steels have been referred to as "TRIP" steels (transformation induced plasticity). All of these transform to martensite during straining. The exact beneficial effect of the prior processing (different amounts of prestraining at different temperatures) is little known. The increase in the yield strength is probably associated with the high dislocation density produced. Also the prior processing has probably a great influence on the ability of the austenite to transform early during testing, which results in a higher ductility.

The austenite → martensite transformation in metastable austenitic steels has been studied by Bressanelli and Moskowitz.[1] They think that martensite forms in the region of the first necking of the specimen and so increases the necked portion, thereby preventing an early breaking. In the case of "TRIP" steels the plastic region (similar to a Luders' band) transverses the specimen and finally the strain tends to be uniform throughout the gage length.

It is thought that the martensite formed during the onset of necking is the most beneficial to the elongation. The specimen continues to elongate considerably after the initial necked region has expanded to reach the ends of the specimen, as there is still some continuation of the austenite → martensite transformation, until a maximum value of martensite content of about 66% is reached. Then the specimen fails, probably because of the low ductility of the martensite.

In the early stages of straining, the increase in length of the specimen is mainly due to the expansion of the plastic region, analogous to a Luders band and the cross-sectional area is then smaller than the initial cross-sectional area of the specimen.

For large plastic strains, the expansion of the specimen cannot be explained only by the dilatation due to the austenite $\rightarrow$ martensite transformation, as the dilatation is very small ($\Delta V = 0.04$). Some authors[2] think that the shear displacements (shear strain of the transformation along the habit plane, $\gamma = 0.20$) are initially statistically oriented. But under the influence of an uniaxial tension, transformation is forced to take place in strongly preferred directions and consequently a large number of elementary shear displacements take place in the same direction giving an important external change in shape.

It is not yet known whether the most important factor in the formation of martensite is the stress or the strain. Angel[2] believes that the governing factor is the supply of mechanical energy through deformation. He feels that stress is of primary importance and that deformation is of some importance too, by its effect on stress conditions, so that local peaks of stress may be formed, activating the transformation.

II. EXPERIMENTAL PROCEDURE

A. Processing of the Alloys

The alloys, with high nitrogen content, were prepared by induction melting in a nitrogen atmosphere and adding iron nitrides. The ingots were forged to approximately 40" × 2-1/2" × 1/2". They were austenitized 1 hr. at 1100°C then cut into smaller pieces. Five pieces were hot-rolled at 900°C to each of the intermediate thicknesses, 0.250", 0.125", 0.0625", in order to produce different amounts of reduction: 80%, 60%, and 20%, by final rolling at different temperatures to the thickness of 0.050". An additional five pieces were rolled from the initial 0.5 in. thickness. This final rolling was preceded by austenitizing all pieces 1 hr. at 1100°C and water quenching. Rolling was then done at different temperatures: RT, 250°C, 350°C, 450°C and 550°C. After rolling, the specimens were air-cooled to room temperature.

Tensile specimens were ground from the 0.050 in. thick sheet. Their dimensions are shown in Fig. 2. Both 1 in. and 5/8 in. specimens were used to test mechanical and magnetic properties.

B. Mechanical Testing

Most of the mechanical testing was done at room temperature and for some specimens in liquid nitrogen on an Instron machine using a cross-head speed of 0.1 cm/minute.

Hardness measurements were made using a Rockwell A scale, in an attempt to follow qualitatively the transformation and to correlate the hardness with the magnetic results.

C. Magnetic Testing of the Specimens

The saturation induction of the specimens was measured during pulling on an Instron machine, using a "permeameter" built specially for this purpose. This equipment is described on Fig. 1. Two detecting coils bucking each other were placed between the poles of an electromagnet. The resulting current from the detecting coils was first integrated, then read on a chart recorder.

The energizing current in the large coils of the electromagnet could be switched continuously from a negative value to a positive value so that the magnetic field between the poles could be reversed from about -6,000 oersted to +6,000 oersted (Fig. 1).

The following operations were made to measure the saturation induction in the specimen:

1. With the specimen removed, search coils A and B were balanced with a divider, to give a minimum signal on the recorder.
2. With the specimen in one of the search coils, any increase, n volts, in the signal on the recorder was due to an additional flux in the specimen.

$$\Delta\phi_{spec} = \Delta(B-H)\ NA_{spec} \doteq B_s NA_{spec}.$$

where

ϕ_{spec}	is the flux in the specimen (Maxwells)
B	is the induction (gauss)
B_s	is the saturation induction
H	is the magnetic field (oersteds)
A_{spec}	is the cross-section area of the specimen.

To calibrate the integrated signal, a square loop flux standard was used; a known variation in the flux $\Delta\phi_{SLFS}$ given by the flux standard gave an integrated voltage read on the recorder.

$$\Delta\phi_{SLFS} \rightarrow n_o \text{ volts on the recorder.}$$

The saturation induction is related to the measured voltage from the integrated signal by:

$$B_s = \frac{1}{2\ NA_{spec}} \times \Delta\phi_{SLFS} \times \frac{n}{n_o} \qquad (1)$$

The factor 1/2 is due to the fact that the magnetic field was switched from a negative value to a positive value. The purpose of this switching was to eliminate any error due to the zero of the induction in the specimen. This zero cannot be attained easily since the remanent magnetization depends on the nature of each specimen.

The signal due to the magnetization of the specimen, n volts, was obtained by subtracting the signal due to an imperfect bucking of the search coils from the total signal read on the recorder.

In Eq. (1) n and n_o are determined by simple reading, $\Delta\phi_{SLFS}$ is known and N is known. Only the area of the specimen, A_{spec}, has to be determined for each magnetic measurement. To determine A_{spec} as a function of the elongation, 2 specimens for each prior treatment were tested. One of them was simply pulled on the Instron machine and its dimensions measured as the pulling proceeded; the other was pulled in position within the permeameter.

As the permeameter could only determine the martensite content in the middle portion of the specimen (only 1/2" of the gage length was in the air gap of the electromagnet), the thickness of all specimens was slightly

decreased (by 0.5 to 2 mils) in the middle, so that the first necking occurred within the poles of the electromagnet.

The B_s determinations were converted to a percentage of martensite by assuming that B_s is proportional to the amount of magnetic phase.

If x is the percentage of martensite, we have:

$$\frac{x}{100} = \frac{B_s}{B_o}$$

B_o is the saturation induction of a completely martensitic specimen. As no completely martensitic specimen could be found for any treatment among the three alloys investigated, B_o was determined using a 4340 steel and an ordinary mild steel.

The results were the following:

for 4340 steel B_o = 19,500 kgauss

for the mild steel B_o = 19,700 kgauss.

A B_o value of 20,000 gauss was arbitrarily chosen to simplify the calculations.

An attempt to calibrate the measurements by x-ray diffraction method was unsuccessful. Most of the difficulty was due to the large amount of cold work in the material. Even metallographic techniques could not be used to resolve the different phases.

To check if the cold work itself affected the magnetic results of a perfect BCC phase, different sheets of mild steel were cold-rolled various amounts. The results show that B_s is proportional to the thickness of the specimens, for a given width. These results suggest that the effect of cold work per se can be neglected.

For one of the steels investigated, Alloy II, both 1" and 5/8" gage length specimens were magnetically tested.

To check the eventual effect on the martensite content of the cold rolling induced by milling, some specimens were measured in the unsectioned condition, and remeasured after having been cut. No significant difference in the B_s value was observed and it was concluded that this effect could be neglected.

III. EXPERIMENTAL RESULTS AND DISCUSSION

A. Introduction

1. Choice of the Alloys

The alloy contents are given in Table I. These were chosen to be similar to some TRIP alloys already tested in this laboratory,[3,4] but in the present study part of the carbon was replaced by nitrogen. The nitrogen solubility was estimated by calculations based on free energy.[5]

These alloys have a very low M_S ($M_S < -196°C$ for Alloy II) and a M_d above room temperature. Nitrogen is known for its influence in giving an austenitic structure and a good combination of strength and ductility.[6,7]

The first alloy (Alloy I) was an Fe-Cr-Mn alloy with 0.217% nitrogen. The manganese was raised to 13% as this element is known, like nickel, to be an austenite stabilizer. An attempt to introduce vanadium in this type of steel was made in order to increase the nitrogen solubility but even a small amount of this element made the steel brittle.

The mechanical properties (yield and tensile strength, elongation and hardness before and after testing) at room temperature are given in Tables II to VI.

2. Magnetic Testing

Magnetic testing of the alloys was carried out at room temperature. Discontinuous testing was employed (measure of B_S at certain points of the tensile curve) and, for certain specimens, continuous testing was done with the large coil remaining permanently energized during the pulling. However, this continuous testing could not be quantitative because the current has to be switched between a negative and a positive value in order to get a quantitative measurement.

Table I. Alloys content

	Fe	Cr	V	Mn	Ni	Si	Mo	N	C
Alloy I	Balance	8	-	13	-	-	-	0.21	-
Alloy II	Balance	9	-	2	8	2	3	0.093	0.25
Alloy III	Balance	9	0.5	2	8	2	1	0.070	0.4

B. Results of the Magnetic Testing

Figures 7, 8, and 9 show the effect of prior processing treatment on the initial martensite content. The magnetic technique used could not discriminate between ferrite and martensite; it is possible that some ferrite was present initially in the specimens. Some ferromagnetic phase was found in all specimens even after austenitizing 2 hrs. at 1100°C.

For Alloy II, at a prior deformation temperature of 250°C, the martensite content increases with the amount of deformation, with a maximum for 80% deformation, and then decreases as the amount of deformation goes up to 90%. The reason for this behavior is not yet clear.

Figures 10 and 11 show the percentage of austenite transformed into martensite at 20% tensile strain in the specimen as a function of prior treatment. The amount of austenite transformed for this given strain increases with the amount of prior reduction, but, surprisingly, an attempt to correlate the increase in martensite transformed at 20% tensile strain with the elongation and the yield strength was unsuccessful.

Table II. Mechanical properties of Alloy I

Prior Treatment (Temp. and amount of deformation, °C)	YS (psi)	UTS (psi)	Elongation %	Hardness before testing, R_A	Hardness after testing, R_A
Specimen 5/8" gage length					
RT 10%	101,000	159,000	28	66	72
RT 20%	144,000	187,000	23	69	75
RT 60%	244,000	244,000	6	76	76
250°C 10%	81,000	159,000	49	--	--
250°C 20%	116,000	157,000	45	--	--
250°C 60%	165,000	166,000	40	--	--
350°C 10%	75,500	159,000	46	--	--
350°C 20%	103,000	161,000	46	62	74
350°C 60%	151,000	171,000	46	70	75
450°C 10%	70,000	156,000	50	54	73
450°C 20%	101,000	159,000	48	64	73
450°C 60%	139,000	170,000	46	68	75
550°C 10%	73,700	152,000	46	60	75
550°C 20%	95,000	156,000	60	55	76
550°C 60%	125,000	175,000	46	62	75
Specimens 1" gage length					
250°C 80%	187,000	187,000	37	73	76.5
250°C 89%	200,000	200,000	22	74	77
350°C 80%	166,000	180,000	43	72	74.5
350°C 89%	171,000	179,000	37	72.5	75
450°C 80%	152,000	180,000	44	71	75
450°C 89%	172,000	187,000	42	72	76
550°C 80%	141,000	177,000	44	70	75
550°C 89%	149,000	170,000	47	71.5	75

Table III.

Mechanical properties of Alloy II. Specimens 1" gage length.

Prior Treatment (Temp. and amount of deformation, °C)	YS (psi)	UTS (psi)	Elongation %	Hardness before pulling, R_A	Hardness after pulling, R_A
Austenitized 2 hrs. 1100°C	47,000	145,000	100	54	78
RT 20%	153,000	180,000	36	69	76.5
RT 60%	250,000	250,000	28	74	78.5
RT 80%	274,000	274,000	31	77	79
RT 90%	382,000	382,000	0	80	80
250 20%	132,000	154,000	39	68	74
250 60%	206,000	219,000	38	74	79.5
250 80%	275,000	275,000	4	78	78
250 90%	288,000	288,000	12	78	80
450 20%	137,000	163,000	50	69	77
450 60%	230,000	235,000	22	74	79
450 80%	224,000	242,000	34	77	81
450 90%	268,000	268,000	10	77.5	79.5
550 20%	138,000	167,000	32	67	74
550 60%	212,000	228,000	31	74	78.5
550 80%	240,000	260,000	34	76	81.5
550 90%	234,000	260,000	20	76	78

Table IV.

Mechanical properties of Alloy II. Specimens 5/8" gage length.

Prior Treatment (Temp. and amount of reduction, °C)	YS (psi)	UTS (psi)	Elongation %	Hardness before pulling, R_A	Hardness after pulling, R_A
RT 20%	135,000	159,000	45	69	76
RT 60%	240,000	240,000	30	75.5	78
RT 80%	252,000	252,000	28	76	81
250 20%	134,000	156,000	42	67	75
250 60%	200,000	214,000	45	73	79
250 80%	278,000	278,000	4	77	77
250 90%	265,000	265,000	12	77	81
450 20%	126,000	160,000	73	67	76
450 60%	210,000	228,000	31	73.5	79
450 80%	233,000	259,000	33	75	78
450 90%	274,000	282,000	22	78	80
550 20%	127,000	157,000	39	66	73
550 60%	188,000	210,000	28	73	78
550 80%	229,000	244,000	38	75.5	82
550 90%	---	---	--	76	78.5

Table V.

Mechanical properties of Alloy III. Specimens 1" gage length.

Prior Treatment (Temp. and amount of deformation, °C)	YS (psi)	UTS (psi)	Elongation %	Hardness before pulling, R_A	Hardness after pulling, R_A
Austenitized 2 hrs. 1100°C	65,000	115,000	42	54	≈68
RT 20%	168,000	234,000	23	74	79
RT 60%	312,000	312,000	2	79.5	79.5
RT 80%	348,000	348,000	0	80	80
250 20%	145,000	202,000	27	70.5	77.5
250 60%	189,000	256,000	24	75.5	80
250 80%	218,000	268,000	24	77	81
250 90%	266,000	266,000	14	78.5	80
450 20%	147,000	227,000	36	71.5	79
450 60%	185,000	255,000	25	75	80
450 80%	201,000	282,000	22	76.5	80
450 90%	220,000	280,000	17	78.5	81
550 20%	148,000	224,000	36	70	79
550 60%	155,000	270,000	21	75.5	79.5
550 80%	190,000	282,000	20	77	80
550 90%	208,000	280,000	16	78	80

Table VI. Ferromagnetic volume in % before and after testing.

Prior Treatment (° C)	Alloy I Long gage length		Alloy II Short gage length		Alloy II Long gage length		Alloy III Long gage length	
	before	after	before	after	before	after	before	after
Aust. 2 hrs. 1100°C	---	--	----	--	0.17	51	0.34	1.9
RT 20%	---	--	----	36	2.15	20	9.25	52
RT 60%	---	--	13.8	42	14.4	44	45.5	---
RT 80%	---	--	10	47	10.2	46	43.3	---
RT 90%	---	--	----	--	38.6	--	--	53
250 20%	---	--	0.38	30	0.33	25	0.137	50
250 60%	---	--	0.69	51	0.3	52	1	67
250 80%	1.43	53	18.1	--	18.7	26	2.3	70
250 90%	5.4	16	4.1	28	3.41	40	4.75	---
450 20%	---	--	0.58	45	0.18	25	0.34	63
450 60%	---	--	0.35	45	0.48	37	0.56	62
450 80%	0.37	62	0.9	67	0.54	64	0.95	67
450 90%	1.8	65	1.67	50	1.43	42	1.8	---
550 20%	---	--	0.45	14	0.33	11	0.3	60
550 60%	---	--	0.20	40	0.33	40	0.58	74
550 80%	0.4	65	0.625	60	0.3	62	1.85	74
550 90%	0.92	63	----	--	7.5	67	67	70

Figures 12 and 13 try to correlate the elongation with the martensite content prior to testing. For both Alloys I and II there seems to be maximums in elongation at very low martensite content and at 10% martensite content.

Figures 14 to 21 show the distribution of the transformation along the gage length for different amounts of nominal tensile strain. There is a good correlation between hardness and amount of magnetic phase. For the tests run at room temperature, the results show clearly the expansion of the plastic region. This expansion is completed at 10% tensile strain.

The magnetic measurements were made at room temperature for the tests run in liquid nitrogen. The results show that both the austenite transformation and expansion of the plastic region occur much more rapidly at the low testing temperature. At 5% tensile strain, the amount of transformation is uniform along the gage length. Figures 20 and 21 show that some specimens which transform very little at room temperature (less than 5% of magnetic phase) transform much more in liquid nitrogen (more than 45%).

Figures 22 to 27 give some examples of magnetic curves correlated with load-displacement curves. Magnetic measurements were made on the three alloys investigated but only those concerning Alloy II are included in this study. The magnetic results are tabulated in Table VI.

Some continuous magnetic measurements were made to correlate the serrations observed in Fig. 24 with increases in the transformation rate. These measurements show a marked increase in the rate of formation of magnetic phase right after the drop of the load; the reaction then proceeds at the same rate as before the serration.

3. Interpretation of the Magnetic Curve

The magnetic curves shown in Figs. 22 to 27 can be interpreted in terms of the phase transformation. There seems to be a very sharp increase in the transformation right after the yield point, until the elongation reaches about 10 to 12%. Then the martensitic reaction rate seems to level off and increase again from 20% elongation up to failure. This can be interpreted in the following way:

a) The necked region, or "plastic region" or Lüders' band, transforms very rapidly the region through which it passes into x% of martensite then expands. The percentage of martensite in the plastic region remains more or less constant during the passage of the band down the remainder of the specimen.

This is confirmed by the static measurements made on different portions of specimens which have been cut up at various stages of the deformation.

b) When the plastic region reaches the ends of the specimen the percentage of martensite along the gage length seems to be uniform (this again is confirmed by the static measurements mentioned above) and the transformation proceeds at a slower rate.

c) For an elongation around 20-25% the transformation rate starts increasing again more rapidly for some specimens. However, for most specimens, the transformation rate levels off at this stage: there can be a stabilizing effect due to the large value of the strain, preventing any further transformation.

C. Accuracy of the Magnetic Measurements

Besides the error due to the calibration (measure of B_o) which has been mentioned earlier, the main source of error comes from the determination of the area of the specimen.

1) For the "dynamic" measurements this cross-sectional area could not be established with better than 10% accuracy in the very early stages of pulling, because of the uncertainty in the position and dimensions of the plastic region in the gage length. The cross-sectional area could be determined with better accuracy for the higher values of tensile strain.

2) For the "static" measurements (on the pieces cut from the specimen) the magnetic method can measure 0.1% of ferromagnetic volume, and can detect even less.

Other sources of error come from the equipment itself: error of reading on the chart recorder, error due to a change in the bucking of the coils during pulling, etc. They were estimated not to exceed 1%.

IV. CONCLUSIONS

The purpose of this investigation was to study the transformation of austenite to martensite during straining and eventually its effect on the mechanical properties of steels. It has been shown that martensite forms very rapidly after the yield point thereby preventing an early failure (due to necking) of the specimen.

However, it was impossible to establish a direct correlation between the amount of austenite transformed into martensite and the elongation or the ultimate tensile strength. The yield strength also seems to be independent of the amount of martensite present prior to tensile testing. The beneficial effect of the prior processing is better explained in terms of density of dislocations and precipitation of carbides and nitrides during prior deformation.

On the other hand, the amount of martensite present prior to tensile testing seems to play a role in the elongation of the specimens. High ductility was found for very low amounts of initial ferromagnetic volume (less than 0.4%). A good combination of strength (above 200,000 psi) and ductility (around 30% elongation) was also found for some specimens containing initially 10% to 12% martensite.

The austenite $\rightarrow$ martensite reaction occurs much more rapidly at liquid nitrogen temperature than at room temperature and more martensite forms for a given strain.

Future work in this direction should encompass the influence of testing temperature and strain rate on the strain induced transformation and improve the calibration of the method.

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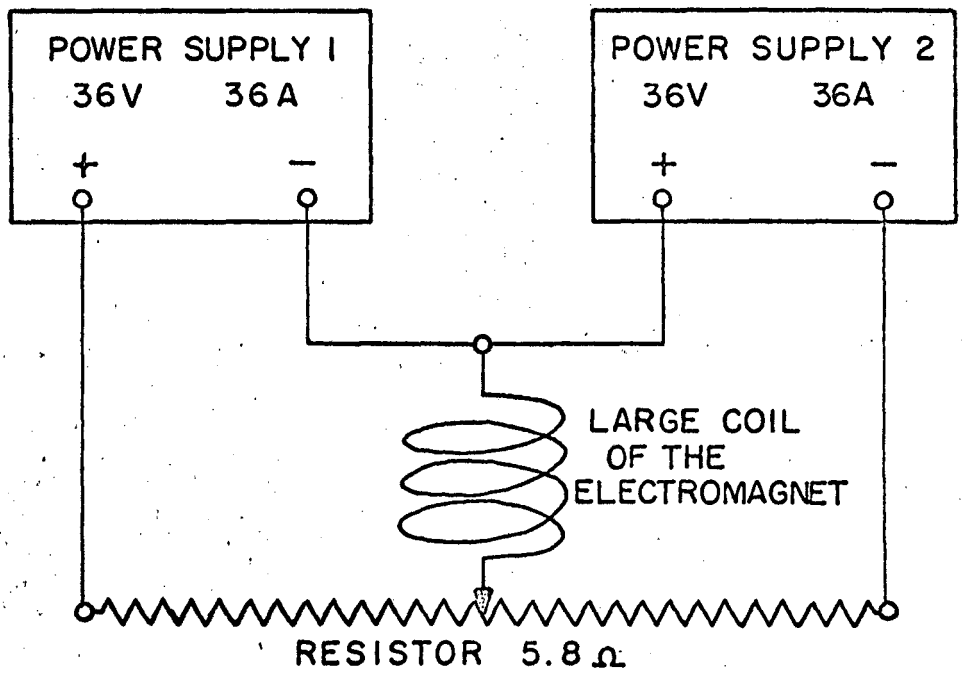
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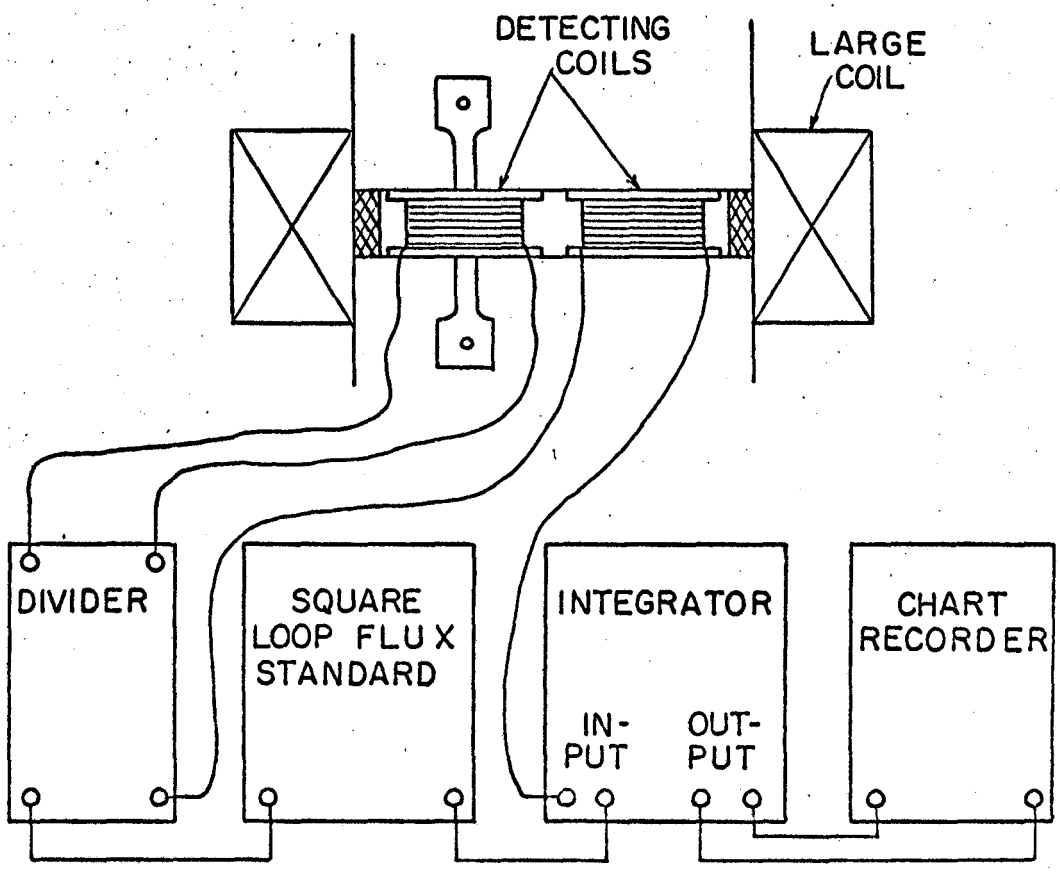
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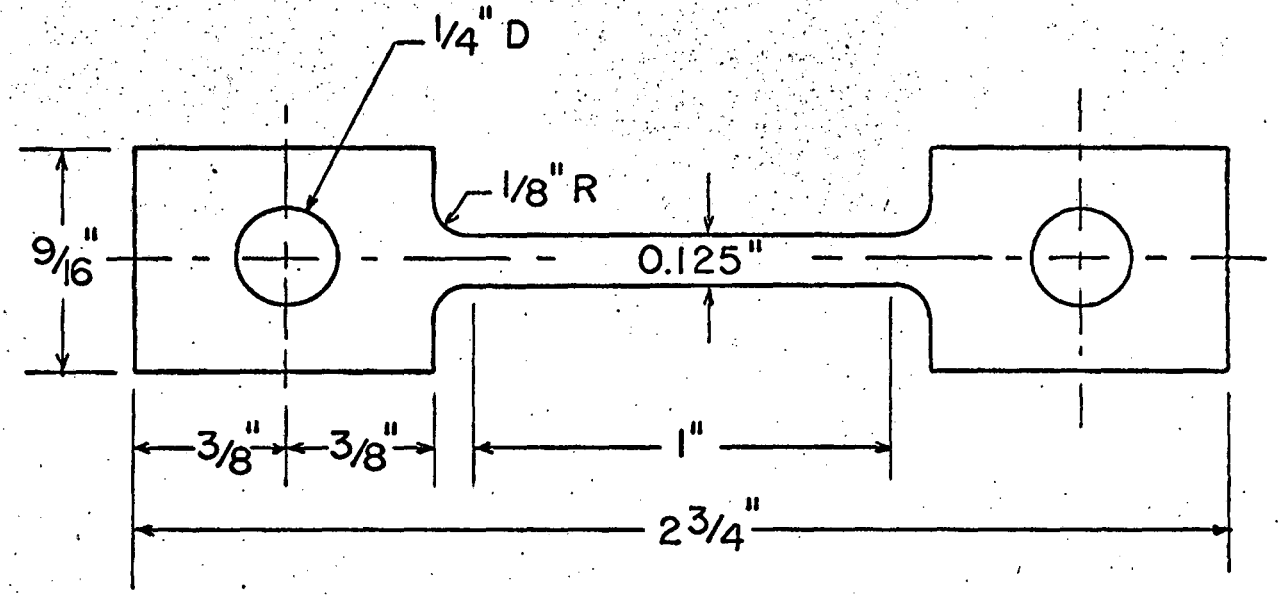
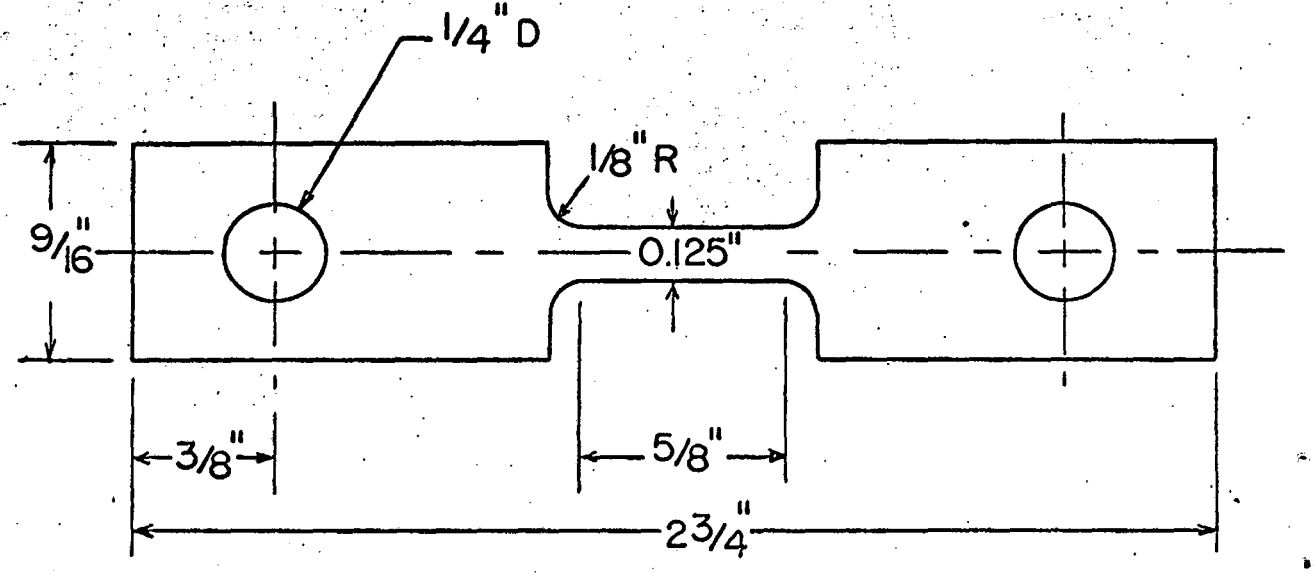
Fig. 1



LONG GAGE TENSILE SPECIMEN.
SCALE: 2" = 1"



SHORT GAGE TENSILE SPECIMEN.
SCALE: 2" = 1"

Fig. 2

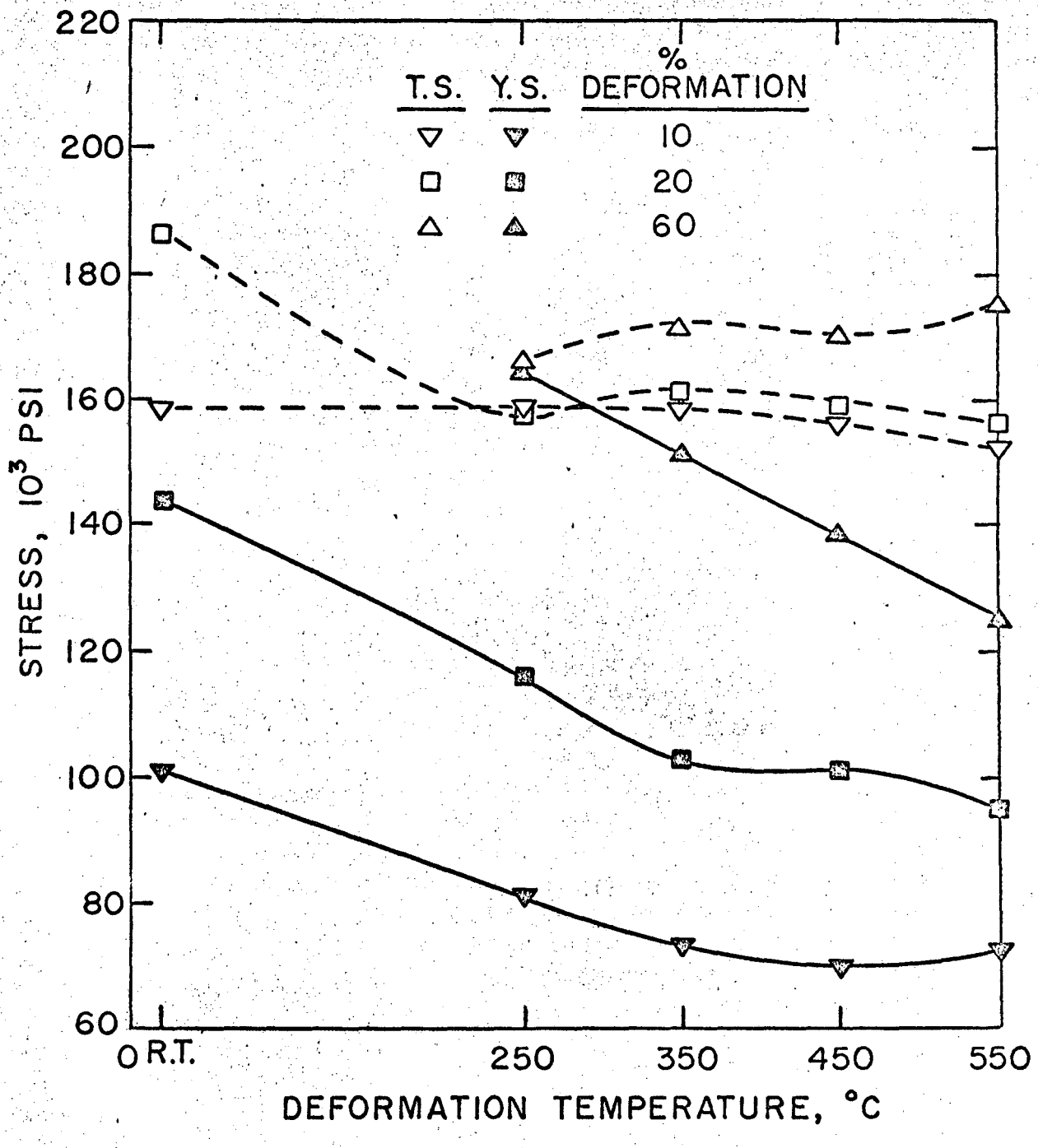

T.S.
Y.S.
% DEFORMATION
10
20
60
220
200
180
160
140
120
100
80
60
STRESS, 10^3 PSI
0 R.T.
250
350
450
550
DEFORMATION TEMPERATURE, °C

Fig. 3

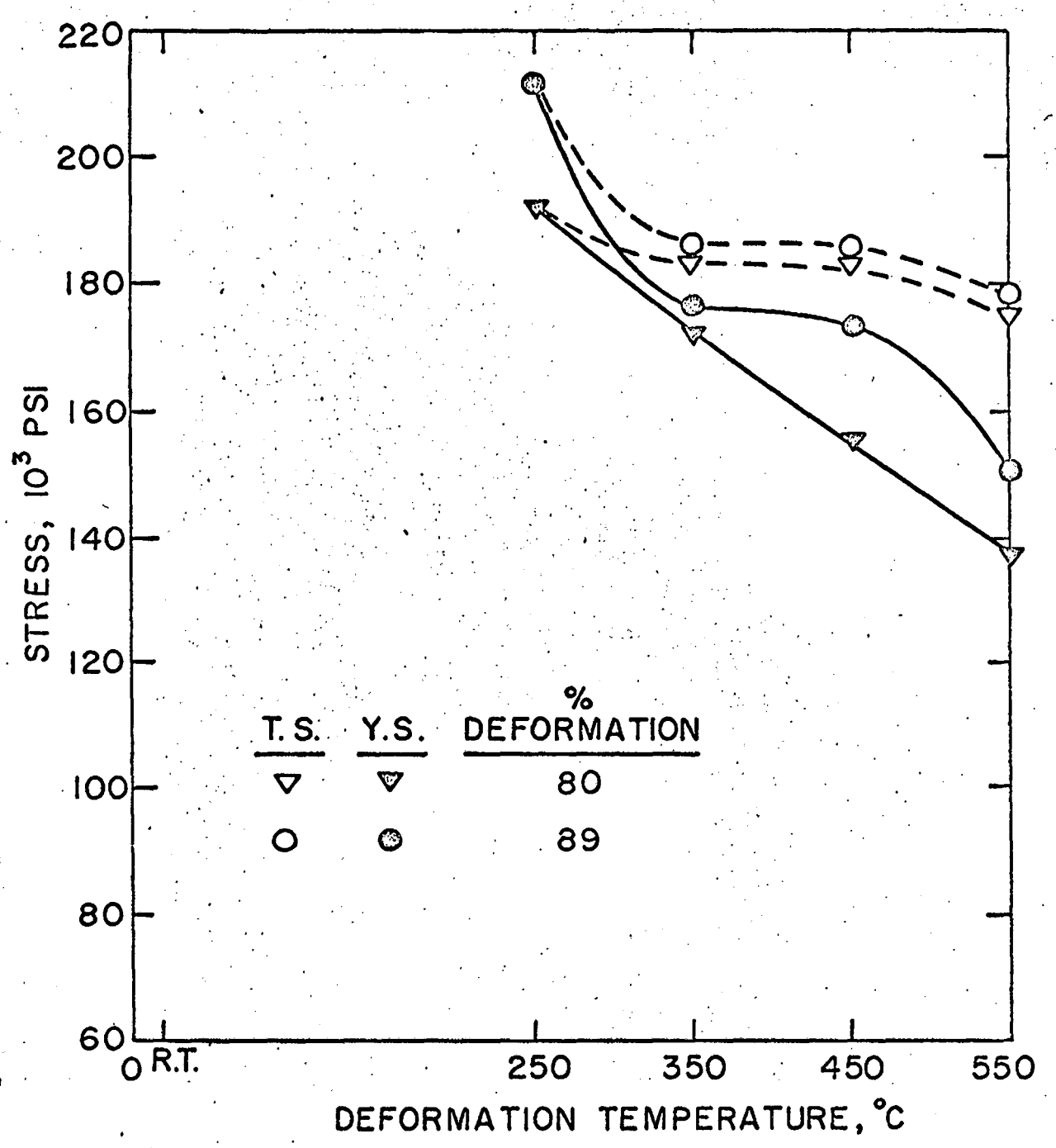

220
200
180
160
140
120
100
80
60
STRESS, 10^3 PSI
0
R.T.
250
350
450
550
DEFORMATION TEMPERATURE, °C
T.S.
Y.S.
% DEFORMATION
80
89

Fig. 4

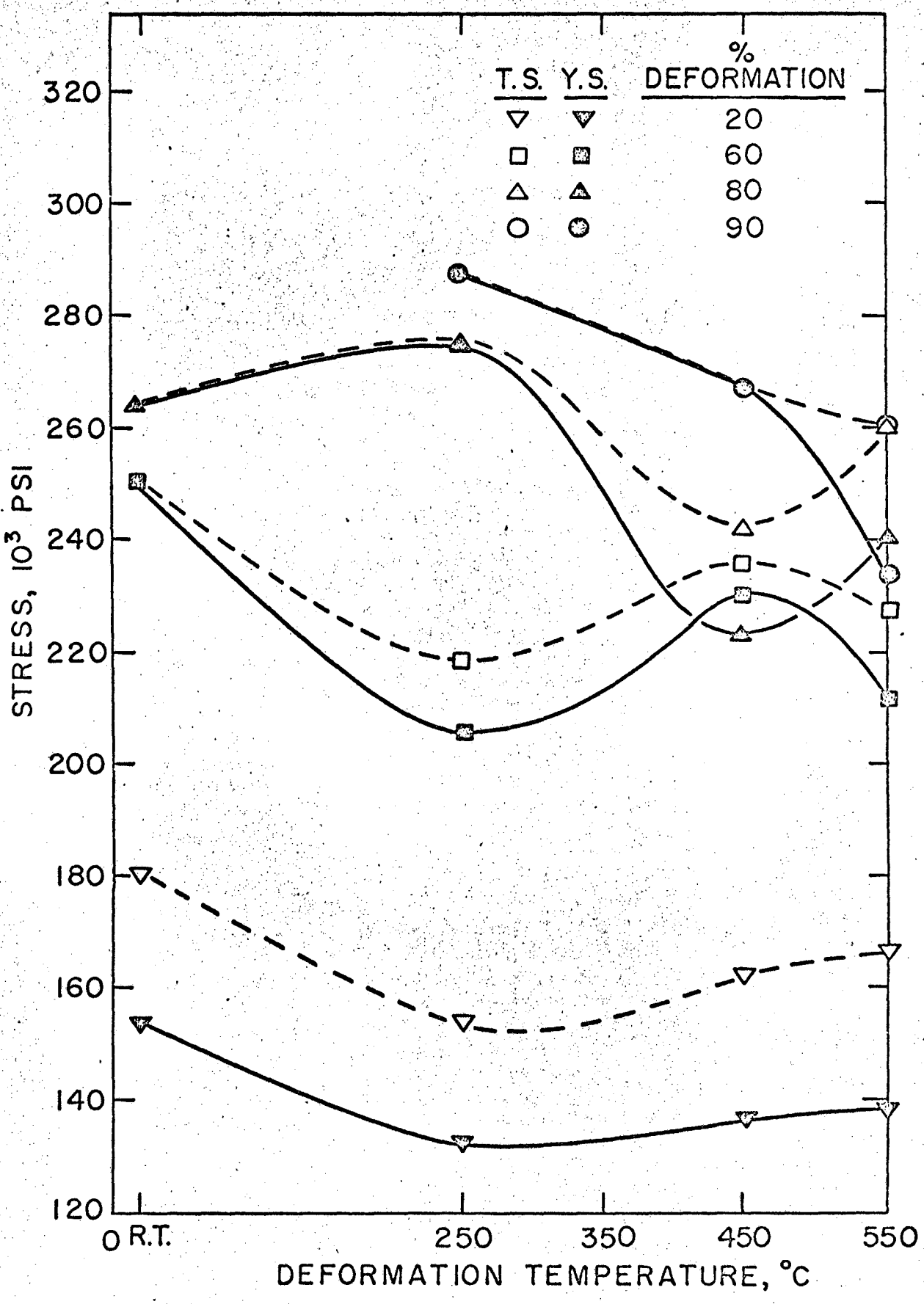

T.S.
Y.S.
% DEFORMATION
20
60
80
90
320
300
280
260
240
220
200
180
160
140
120
STRESS, 10^3 PSI
0
R.T.
250
350
450
550
DEFORMATION TEMPERATURE, °C

Fig. 5

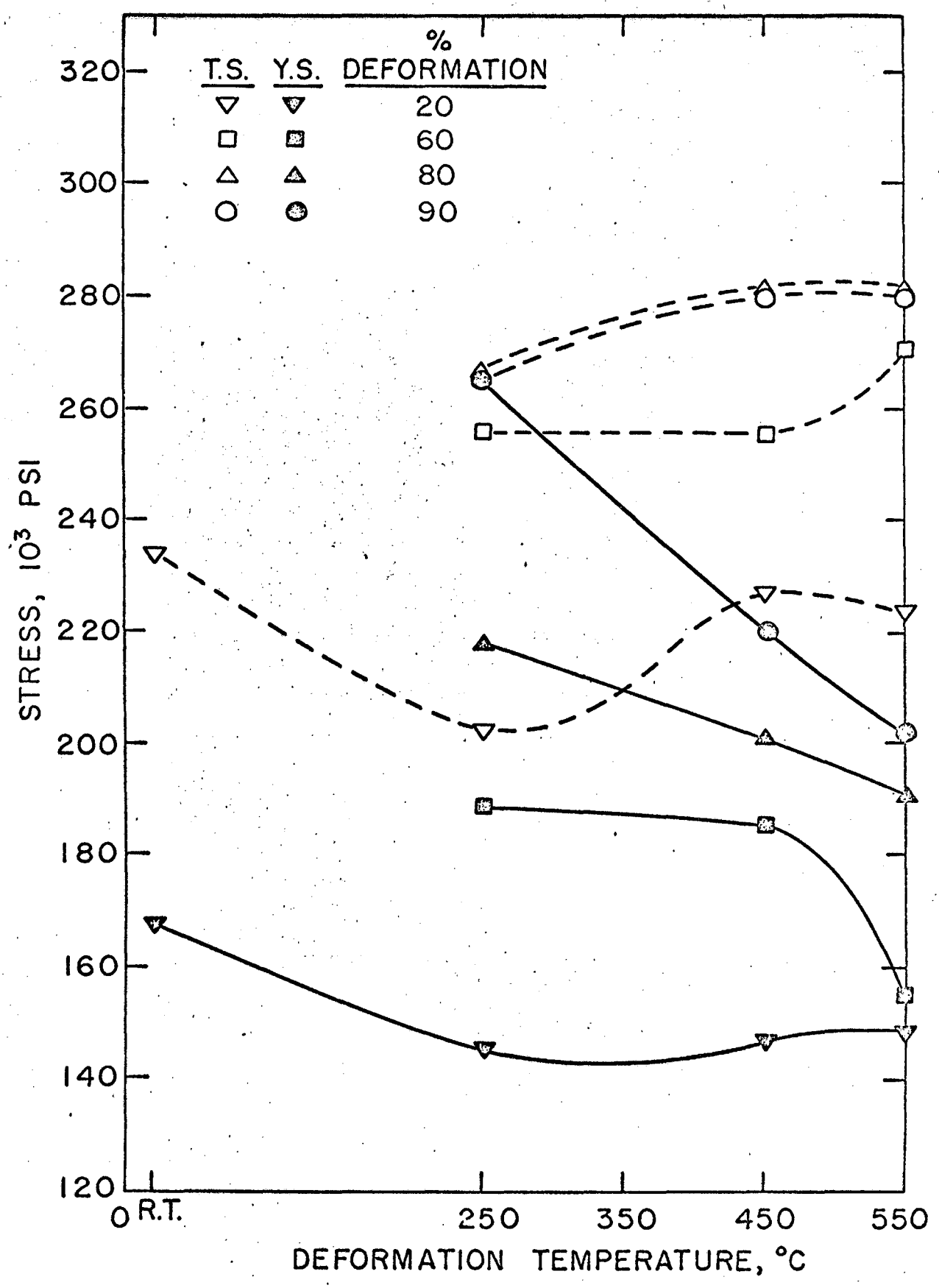

T.S.
Y.S.
% DEFORMATION
20
60
80
90
320
300
280
260
240
220
200
180
160
140
120
STRESS, 10^3 PSI
0
R.T.
250
350
450
550
DEFORMATION TEMPERATURE, °C

Fig. 6

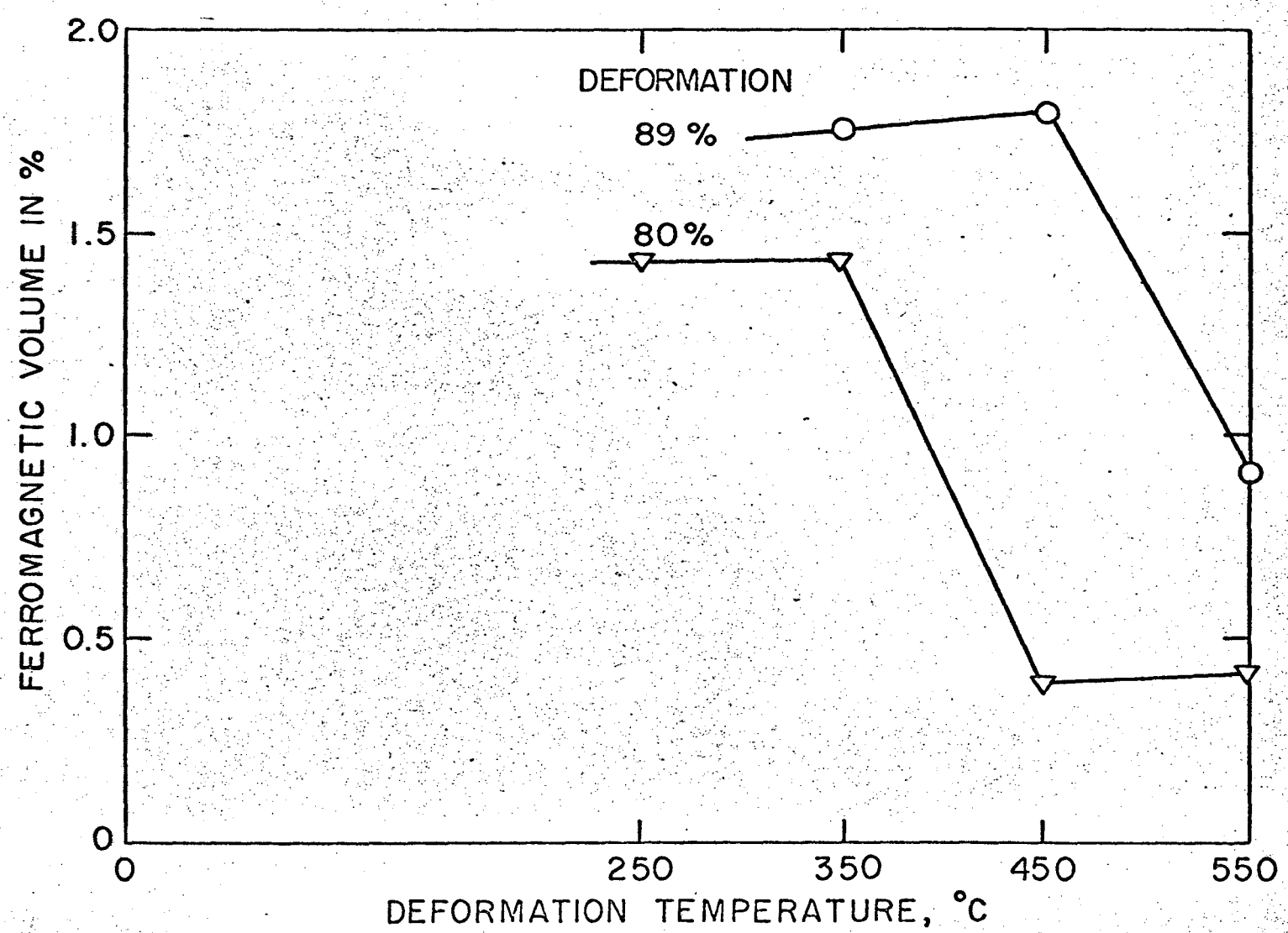


Fig. 7 Alloy I Initial ferromagnetic volume as a function of prior treatment

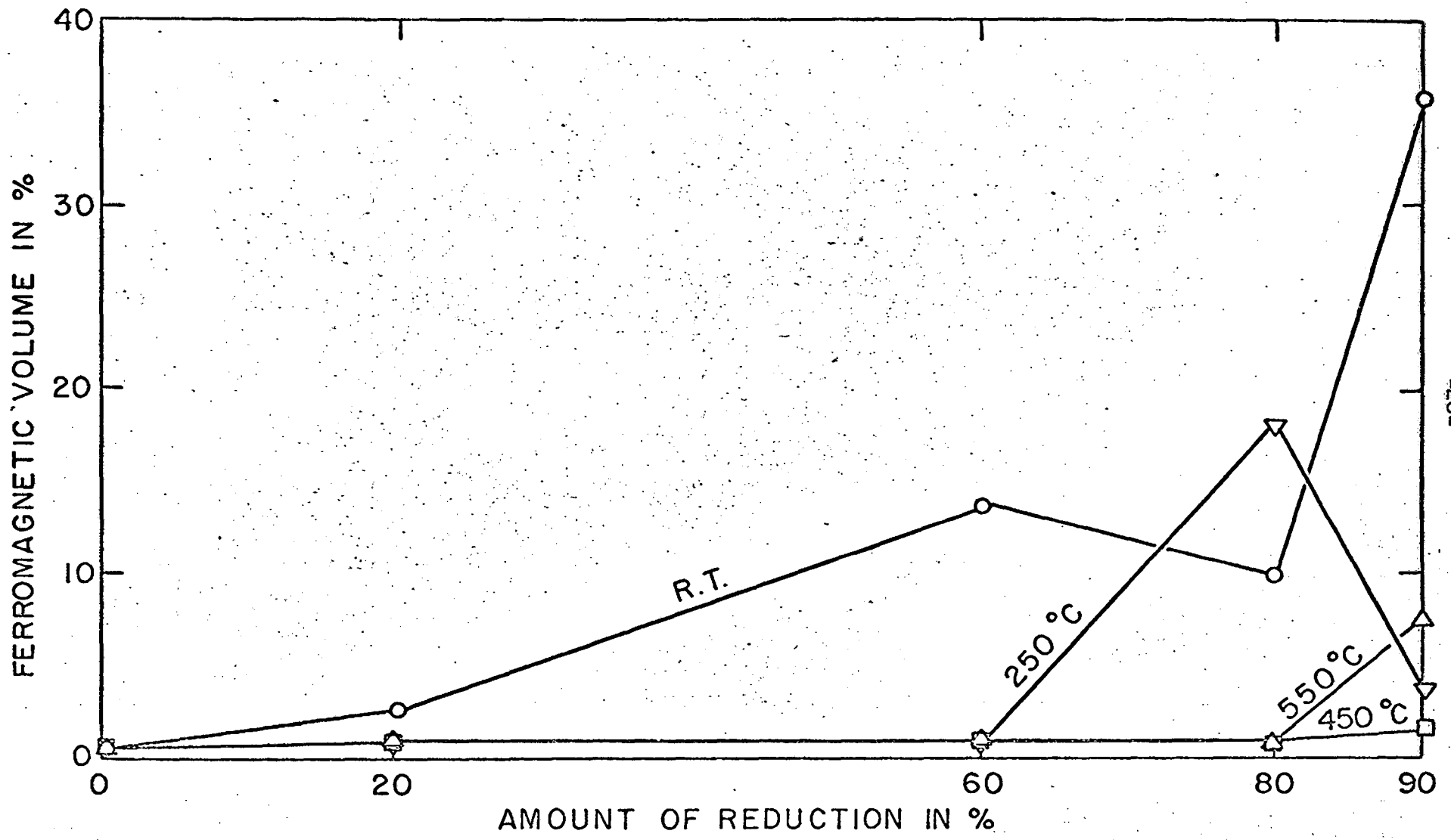


Fig. 8 Alloy II Initial ferromagnetic volume as a function of prior treatment

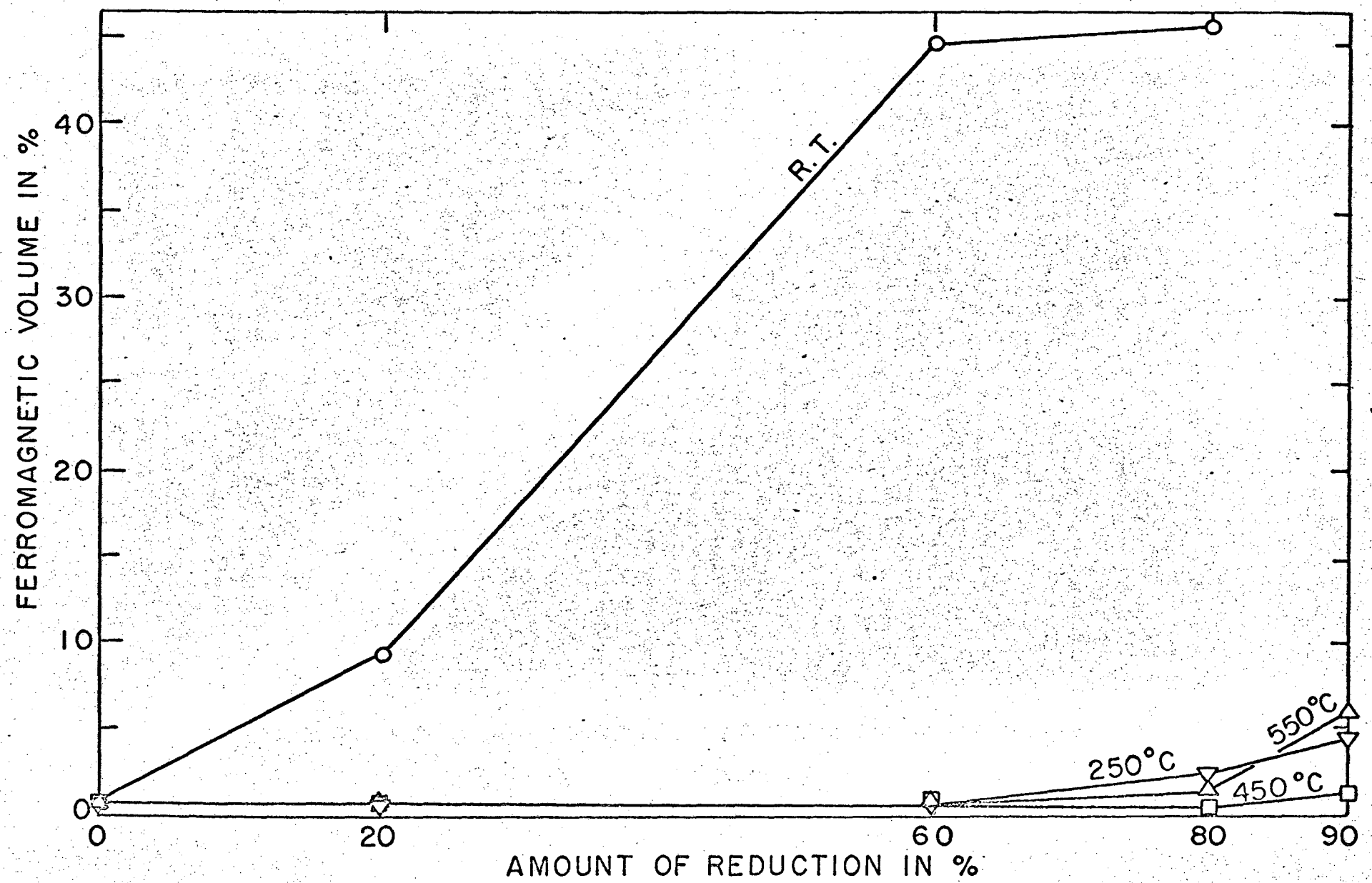


Fig. 9 Alloy III Initial ferromagnetic volume as a function of prior treatment

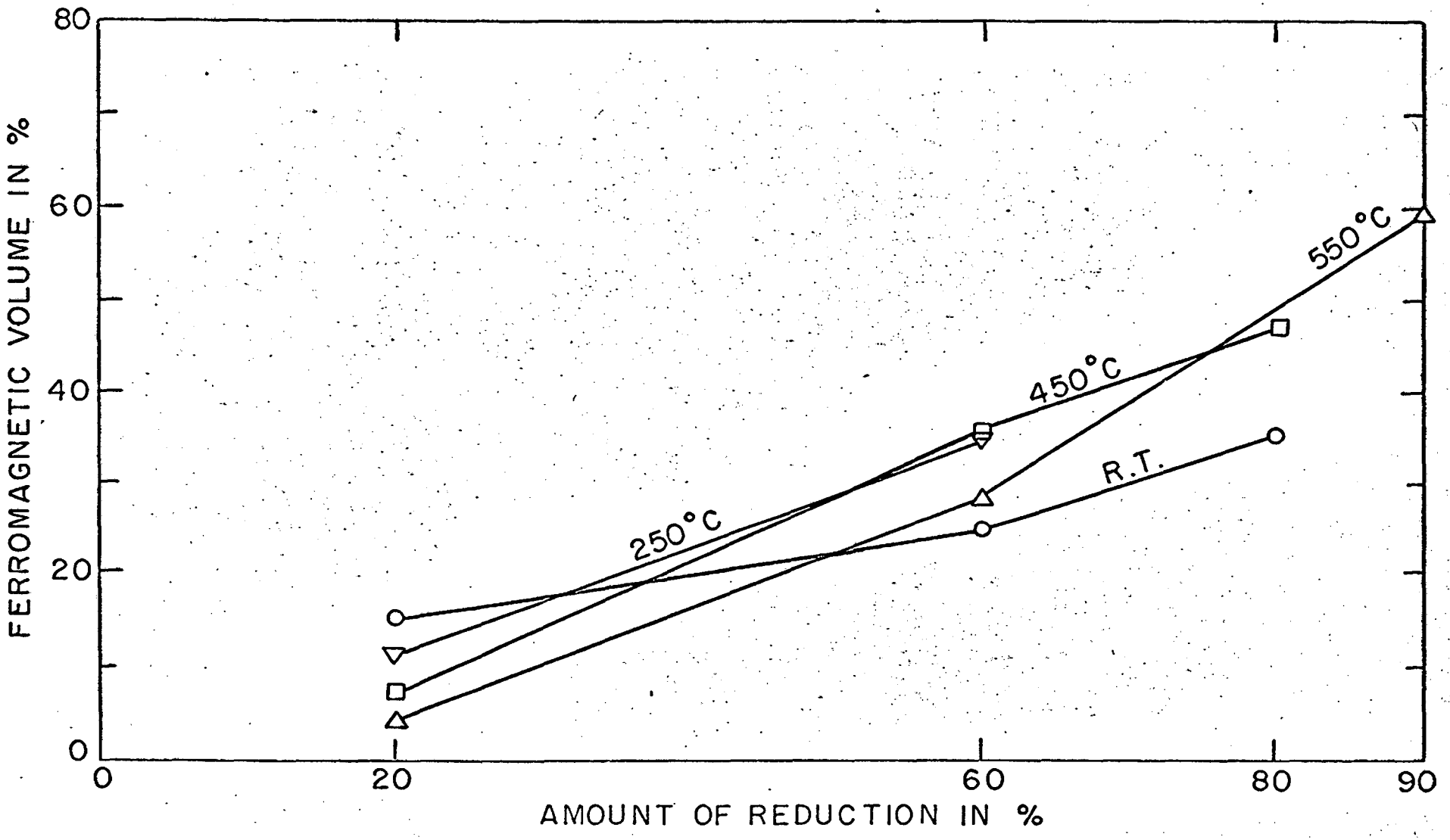


Fig. 10 Alloy II Increase in % of ferromagnetic volume formed at 20% tensile strain

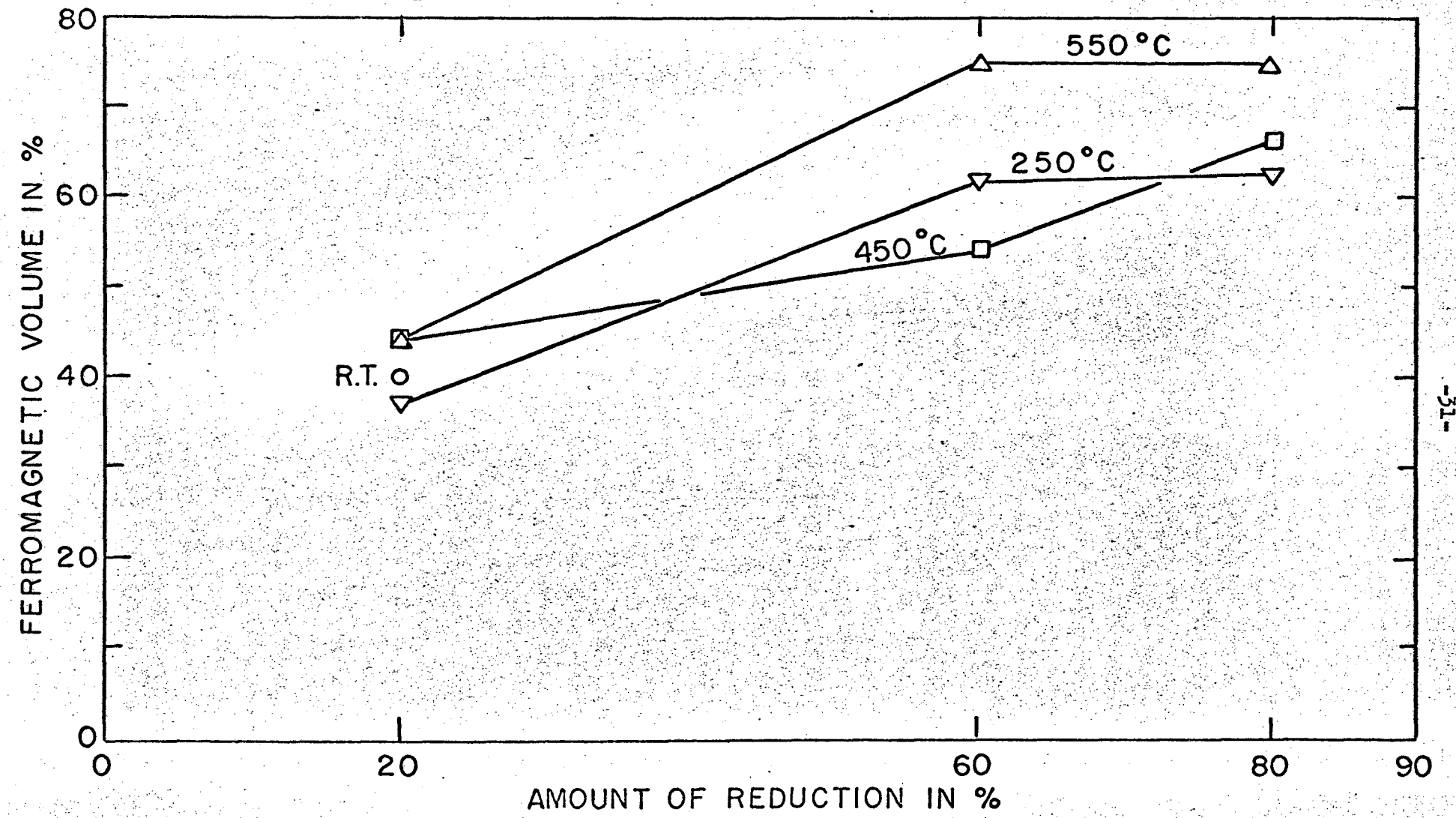


Fig. 11 Alloy III Increase in % of ferromagnetic volume formed at 20% tensile strain

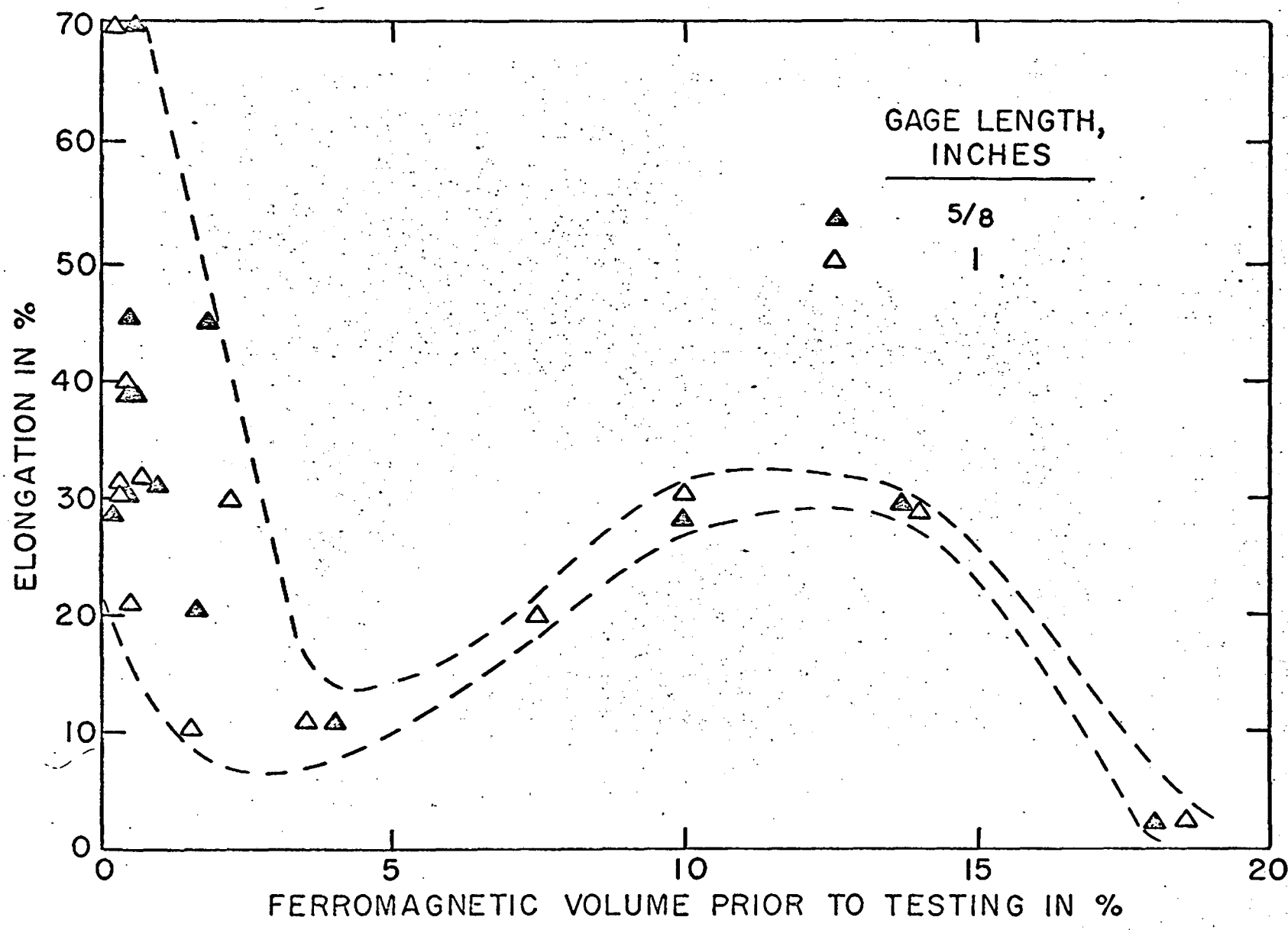


Fig. 12 Alloy II

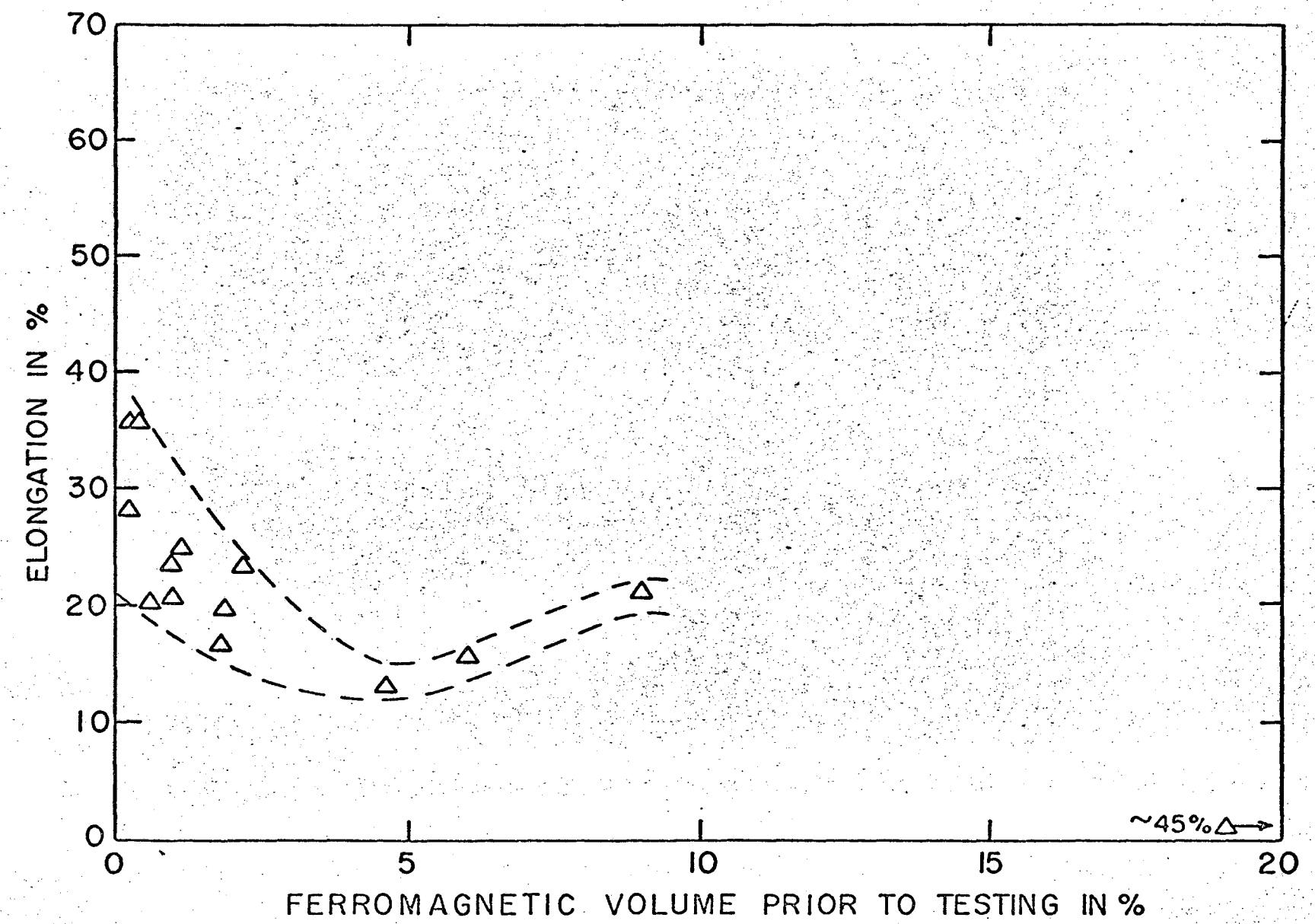


Fig. 13 Alloy III

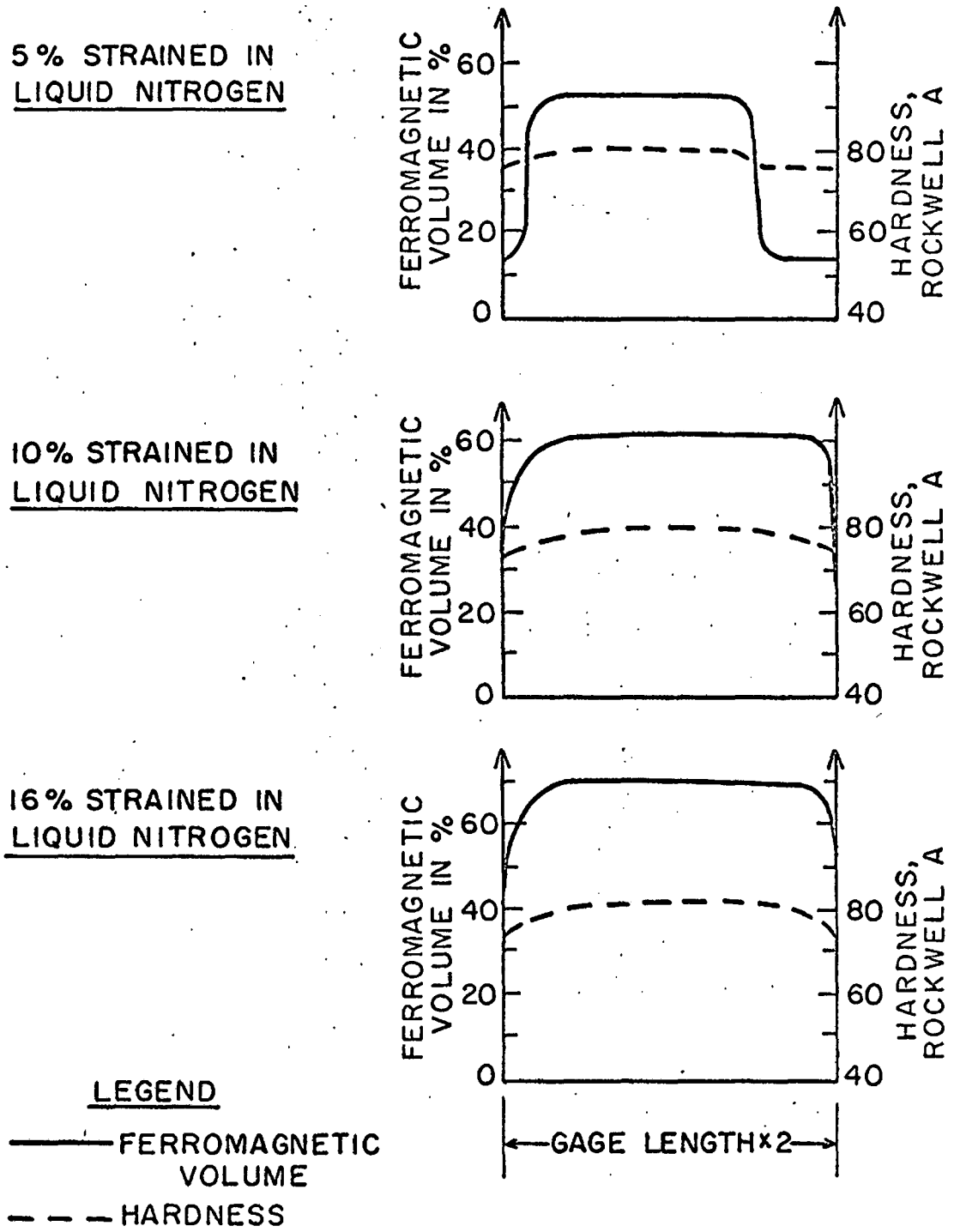


Fig. 14 Alloy II - Distribution of hardness and magnetic phase along the gage length.
Prior treatment RT 60%

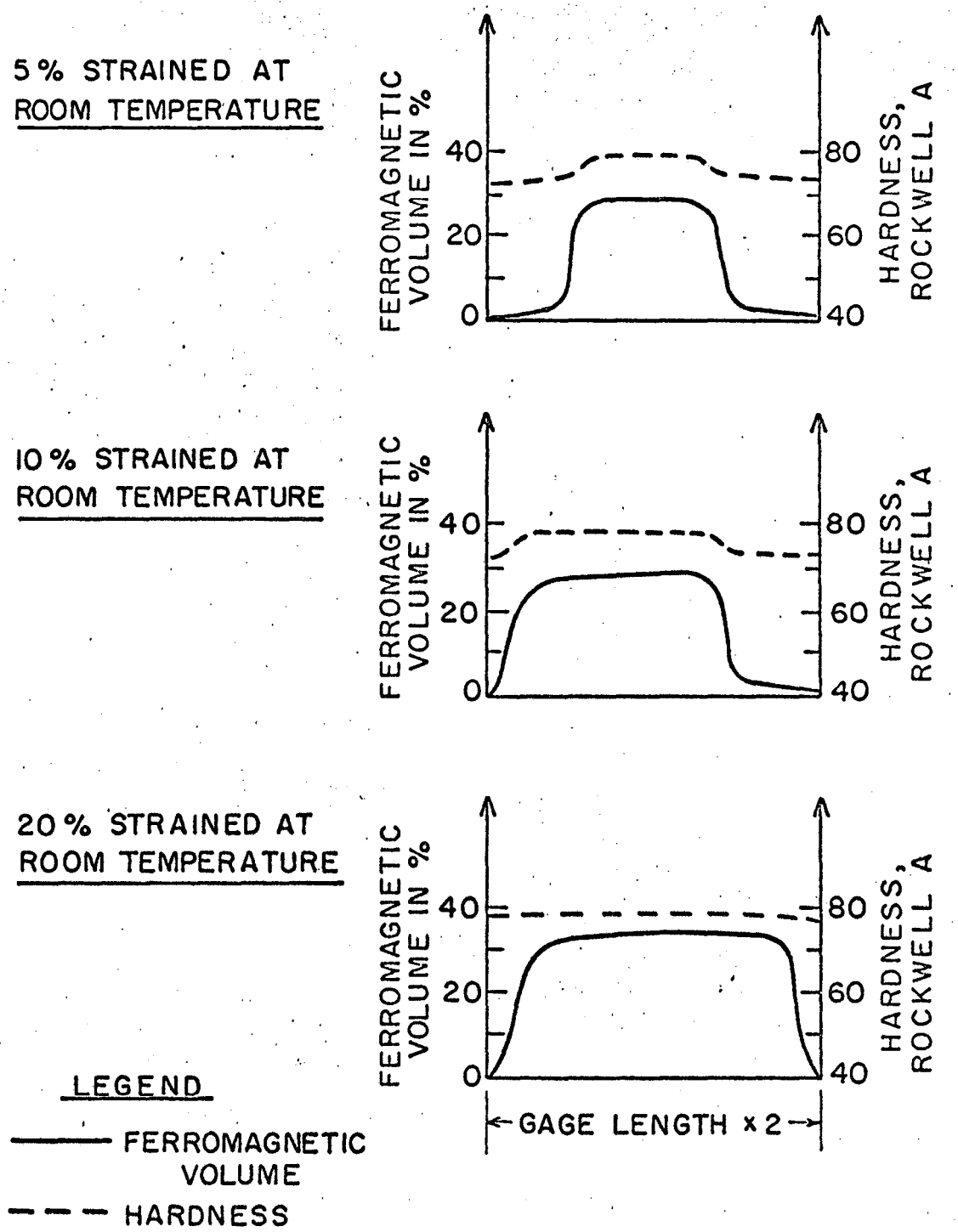


Fig. 15 Alloy II - Distribution of hardness and magnetic phase along the gage length.
Prior treatment 450°C 60%

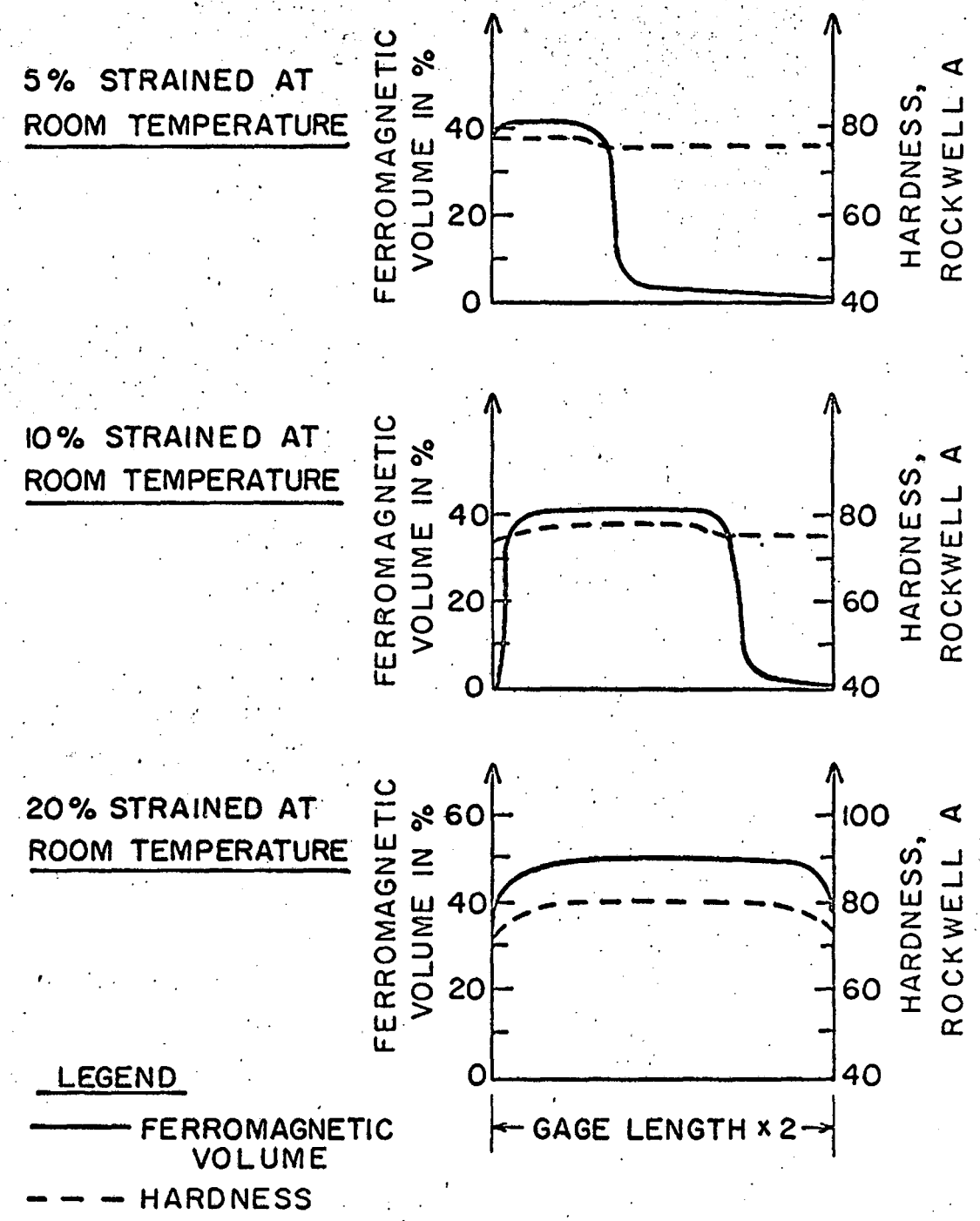


Fig. 16 Alloy II - Distribution of hardness and magnetic phase along the gage length
Prior treatment 450°C 80%

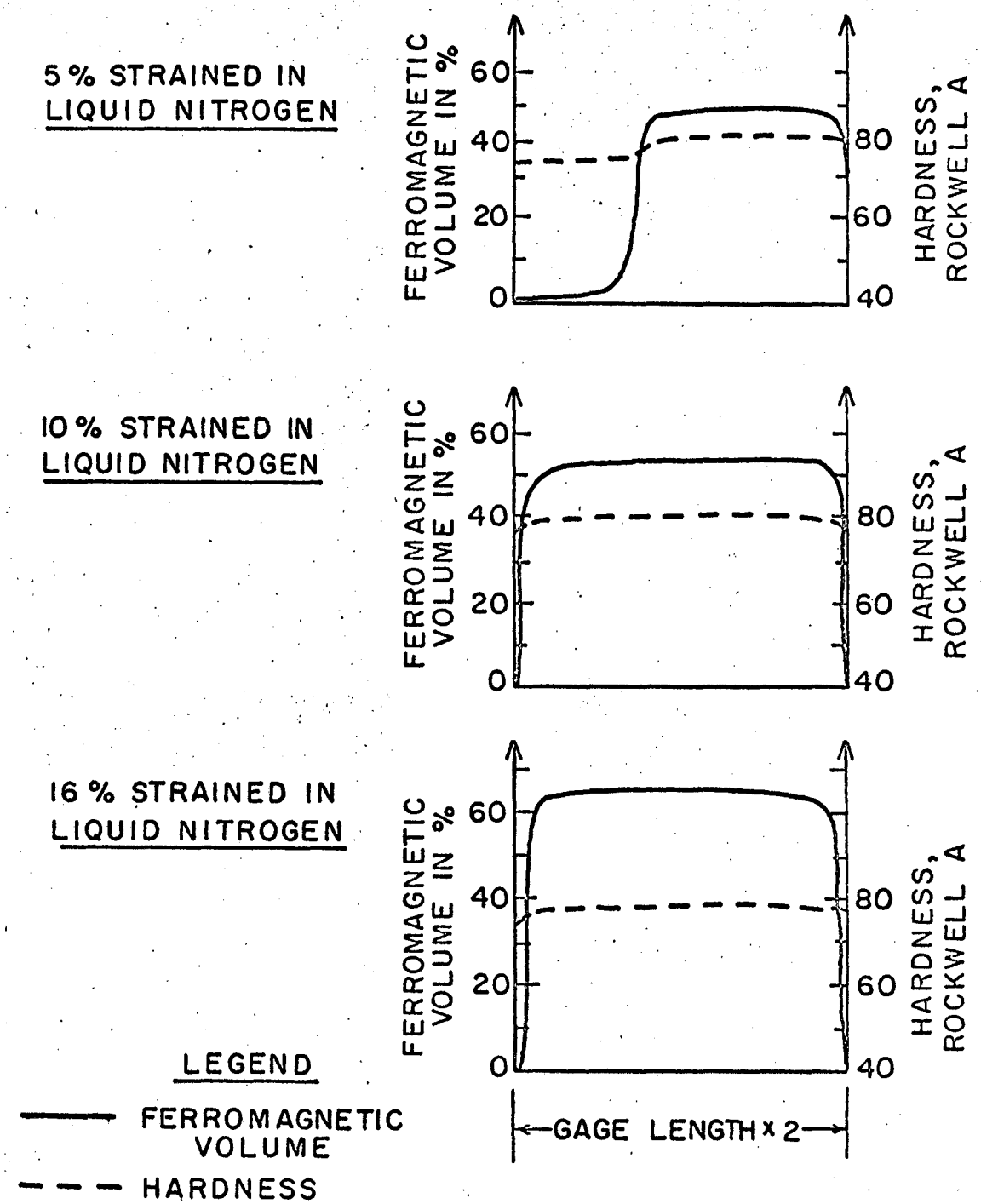


Fig. 17 Alloy II - Distribution of hardness and magnetic phase along the gage length
Prior treatment 550°C 80%

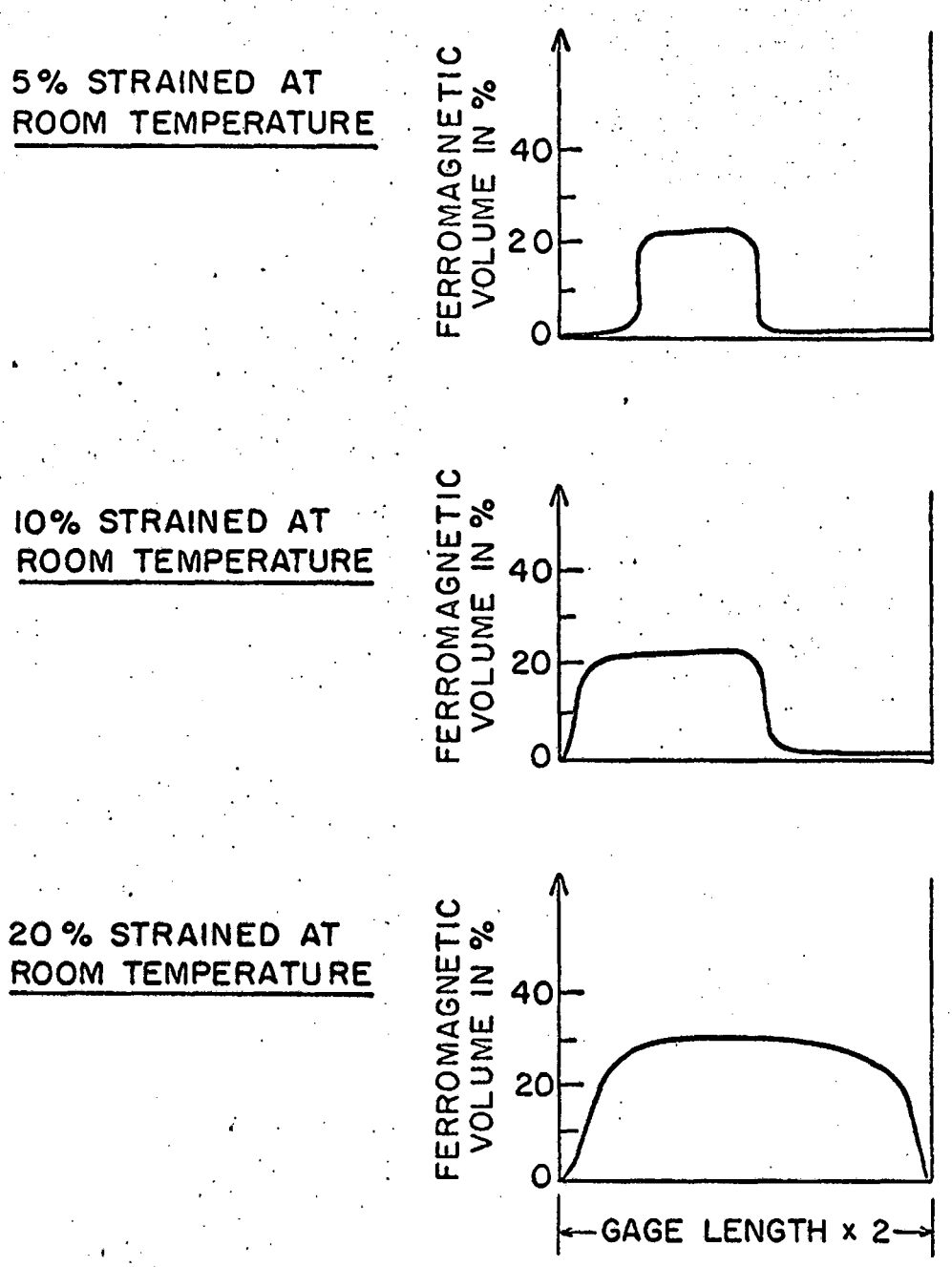


Fig. 18 Alloy II - Distribution of magnetic phase along the gage length
Prior treatment 250°C 60%

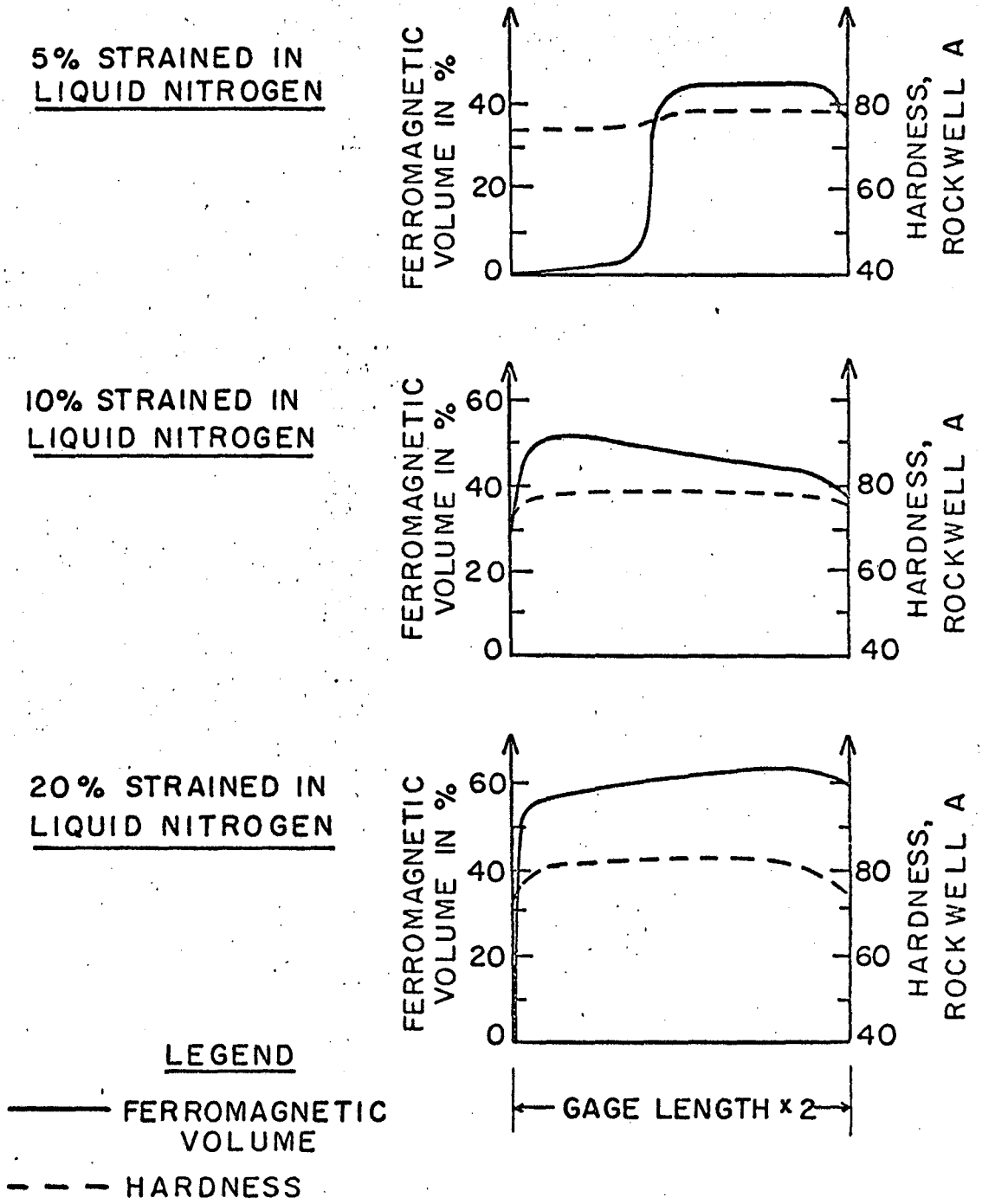


Fig. 19 Alloy II - Distribution of hardness and magnetic phase along the gage length
Prior treatment 250°C 60%

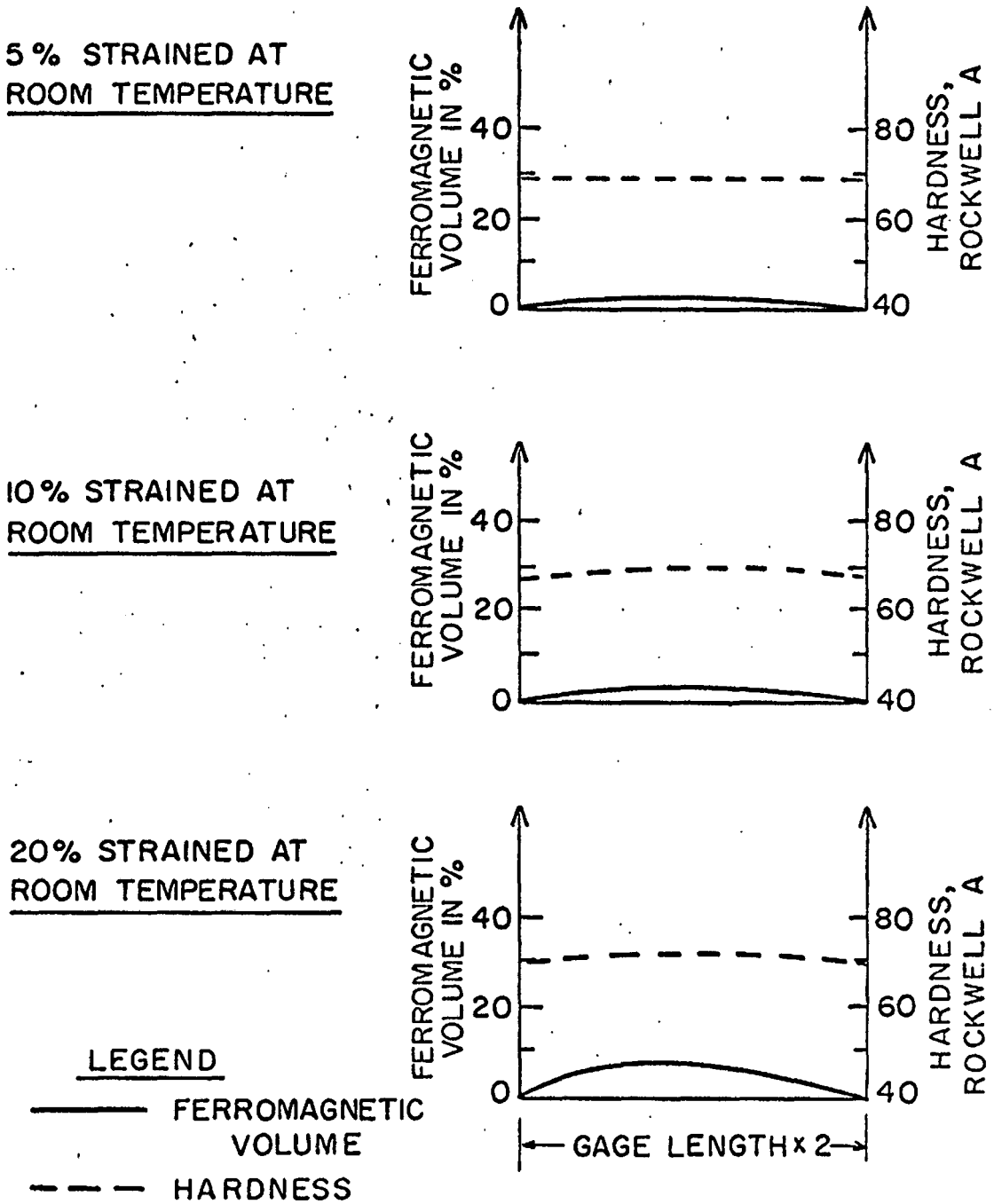


Fig. 20 Alloy II Distribution of hardness and magnetic phase along the gage length
Prior treatment 450°C 20%

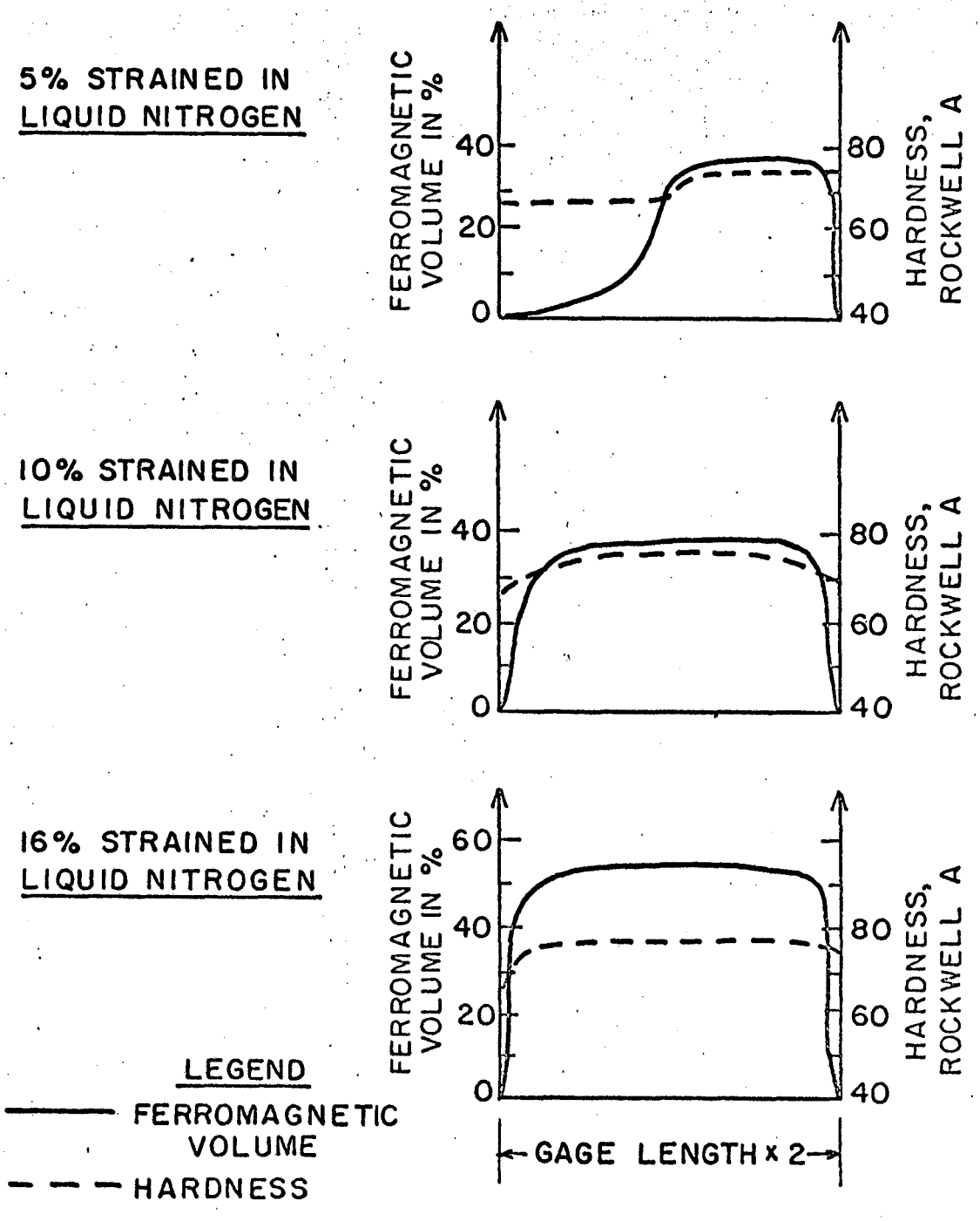


Fig. 21 Alloy II Distribution of hardness and magnetic phase along the gage length
Prior treatment 450°C 20%

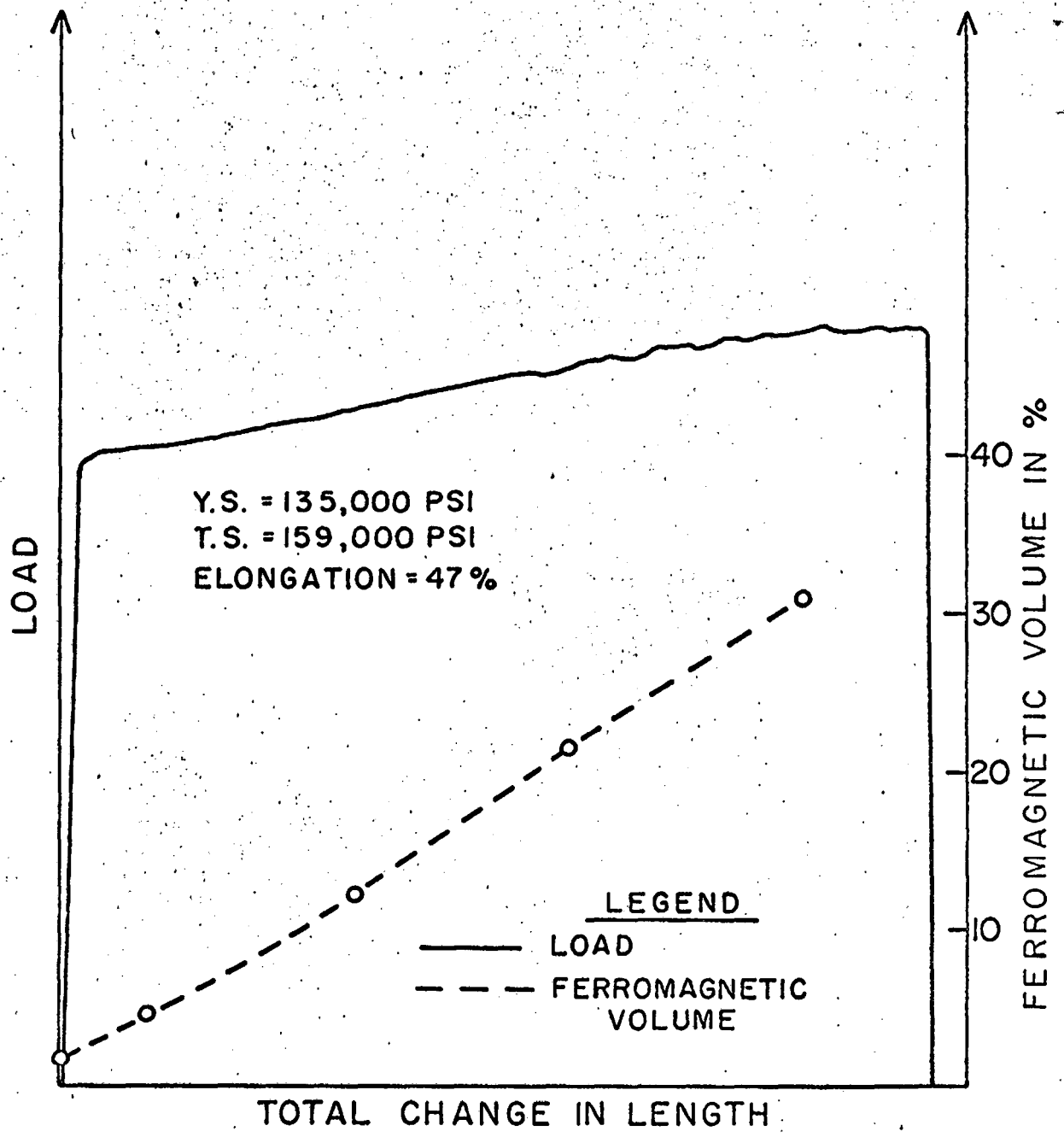


Fig. 22 Alloy II Load displacement curve and corresponding ferromagnetic volumes
Prior treatment RT 20%
Short gage specimen

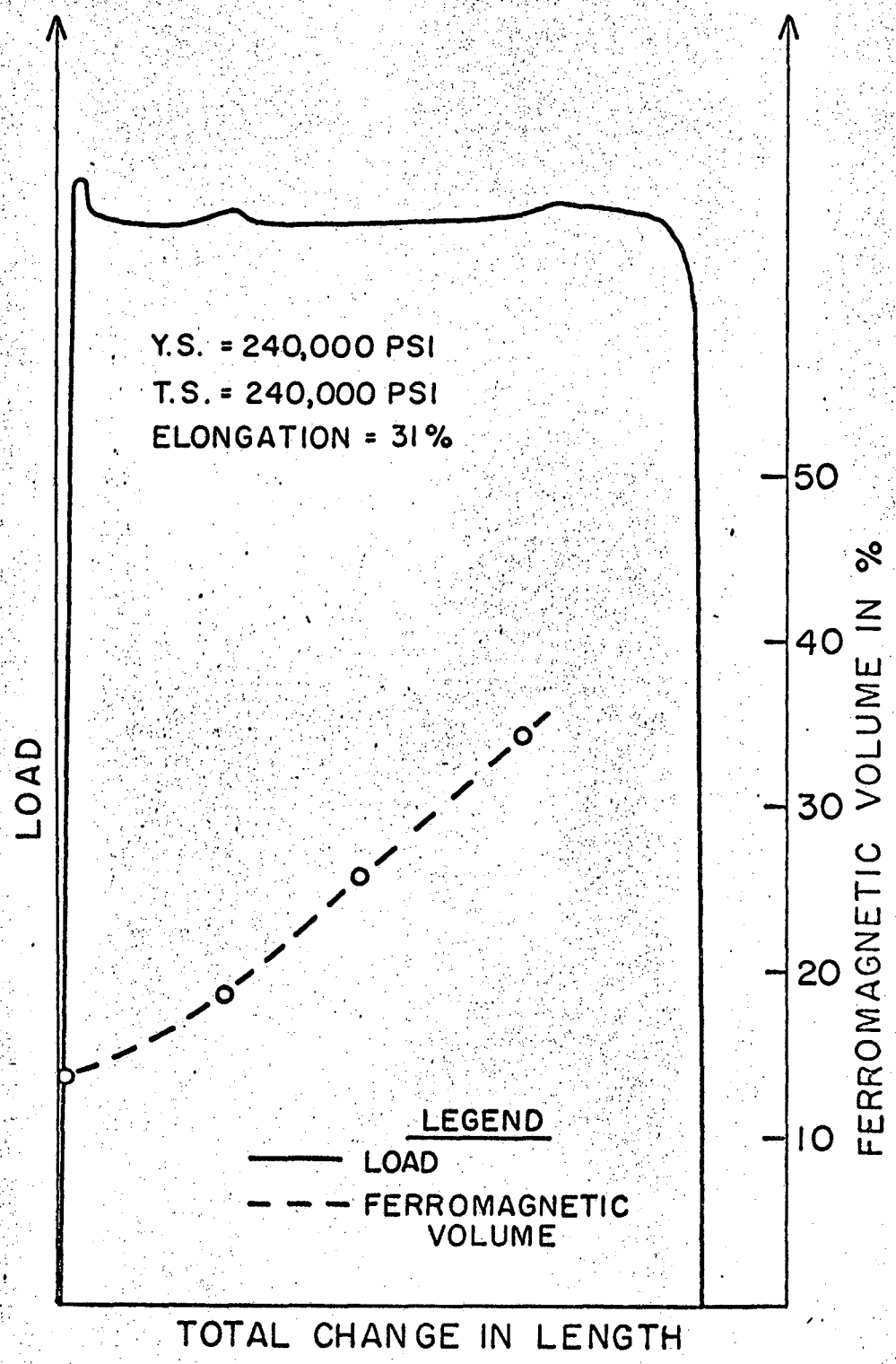


Fig. 23 Alloy II - Load displacement curve and corresponding ferromagnetic volumes

Prior treatment RT 60%
Short gage specimen

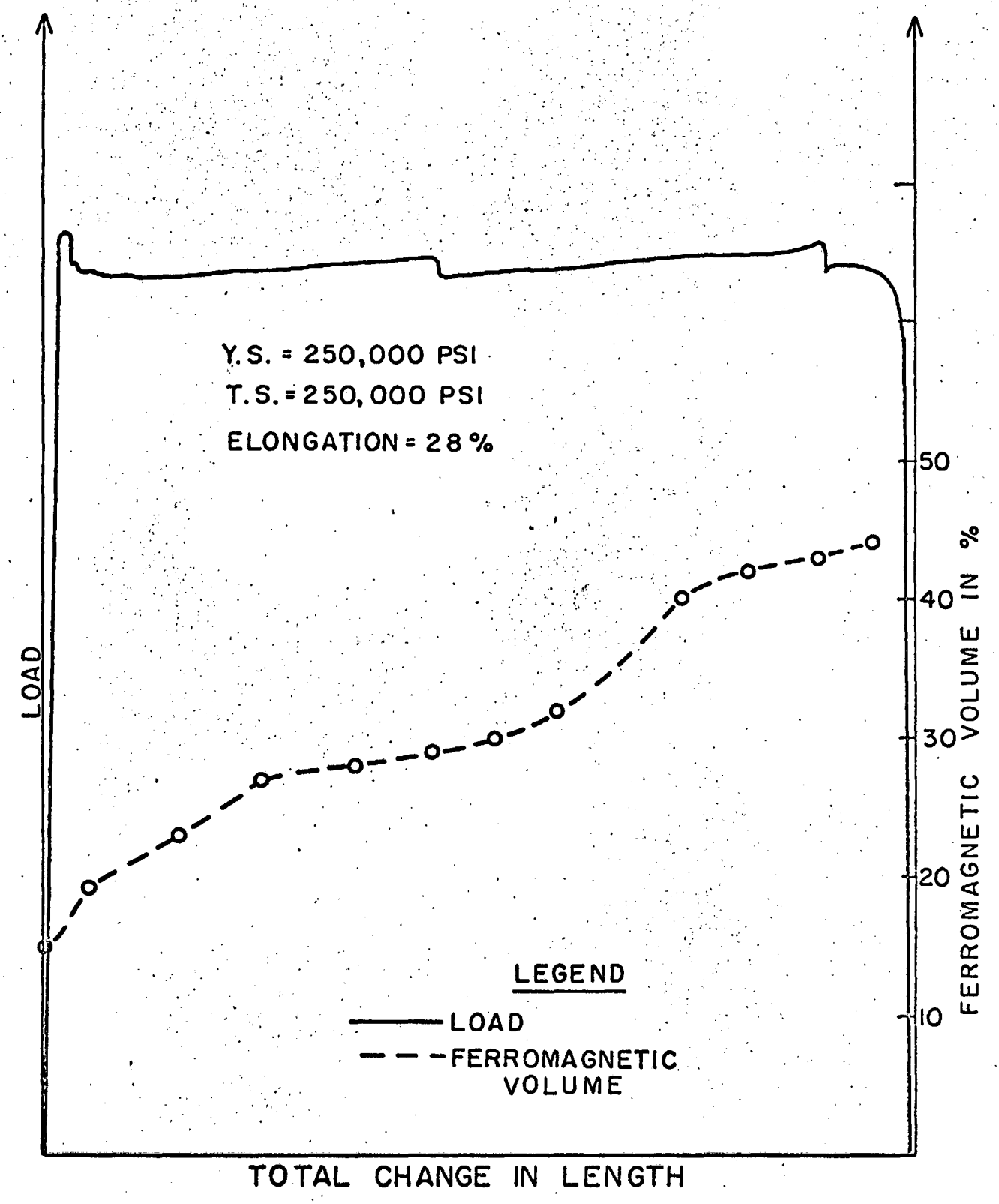


Fig. 24 Alloy II - Load displacement curve and corresponding ferromagnetic volumes
Prior treatment RT 60%
Long gage specimen

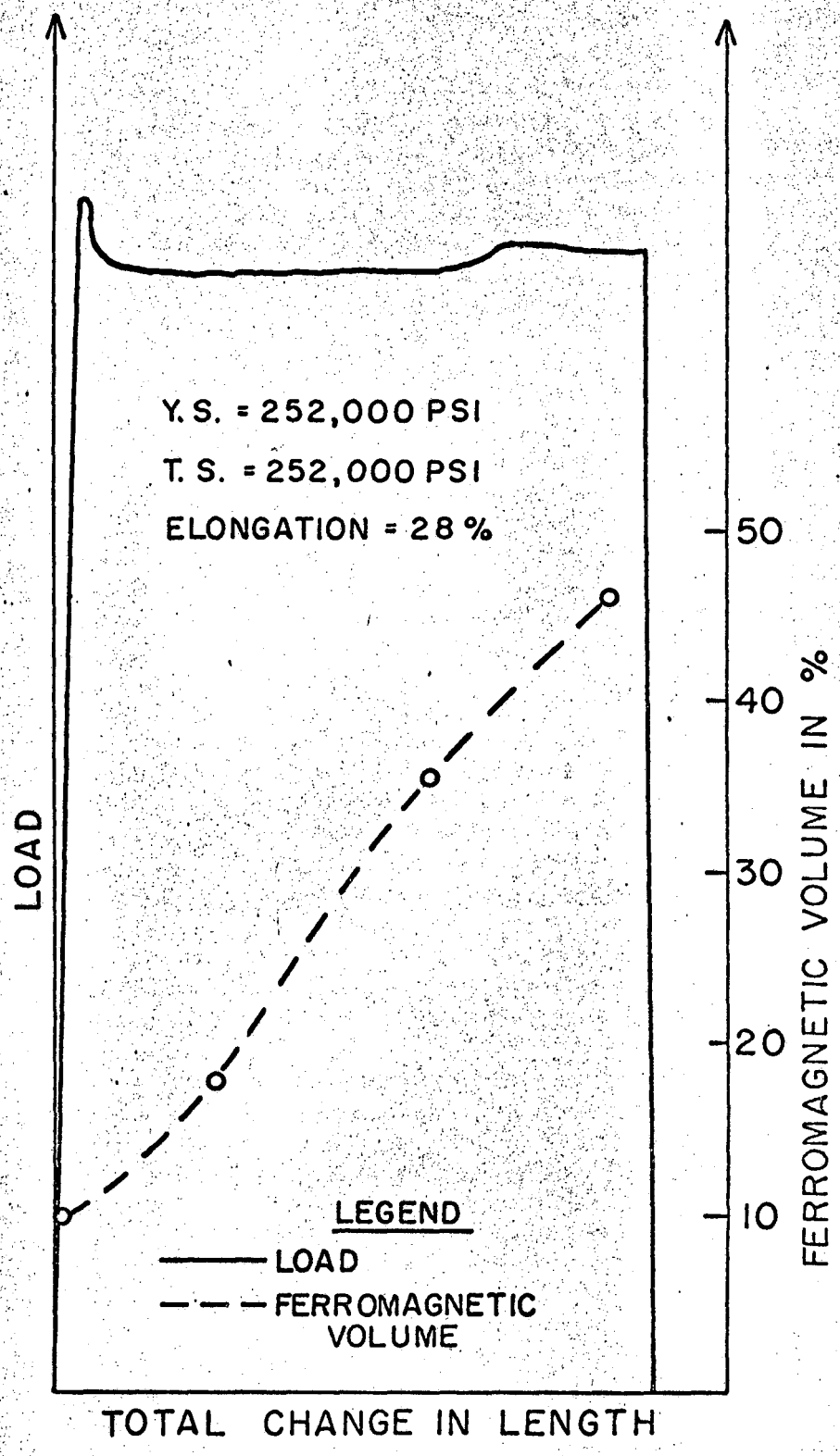


Fig. 25 Alloy II - Load displacement curve and corresponding ferromagnetic volumes
Prior treatment RT 80%
Short gage specimen

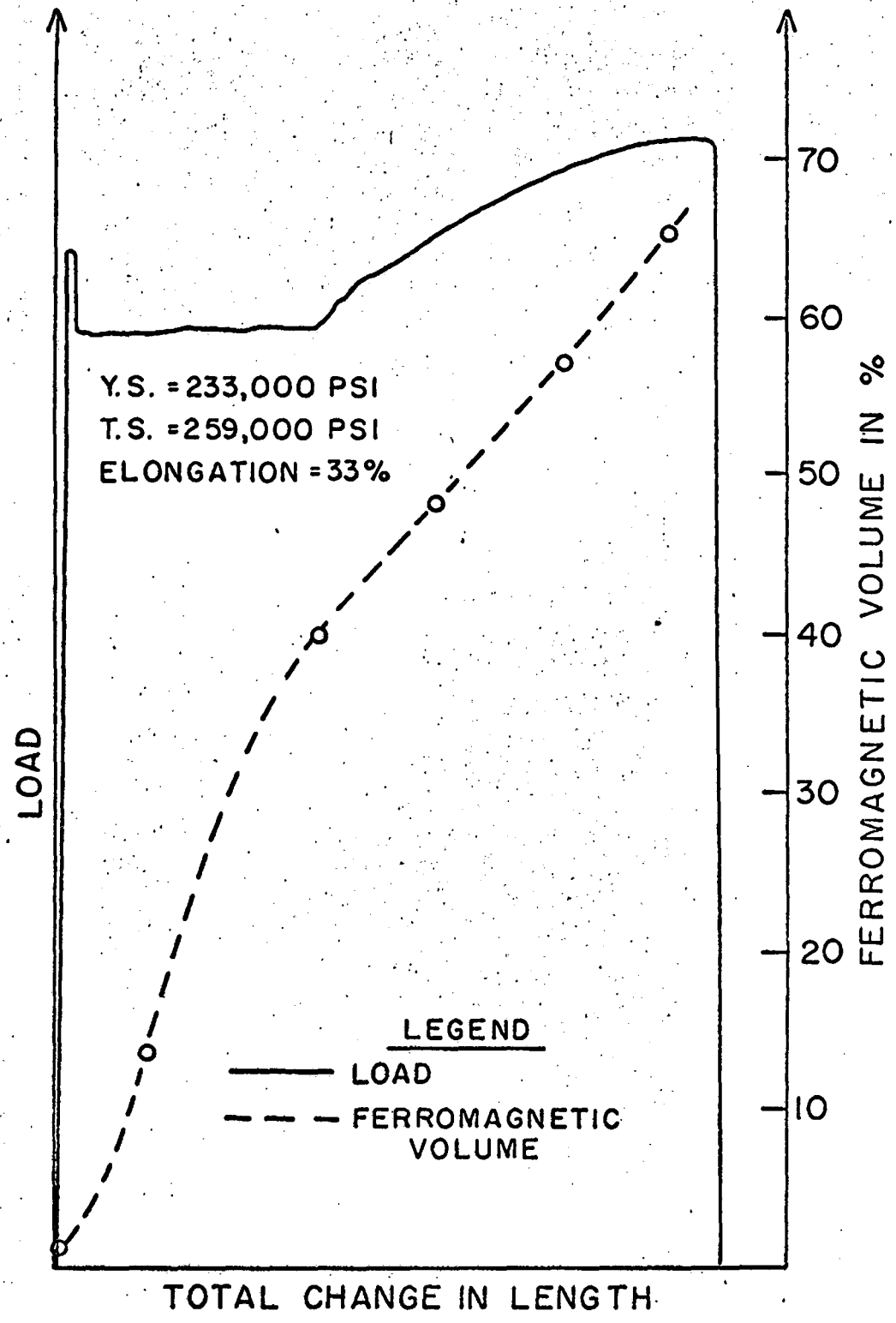


Fig. 26 Alloy II - Load displacement curve and corresponding ferromagnetic volume
Prior treatment 450°C 80%
Short gage specimen

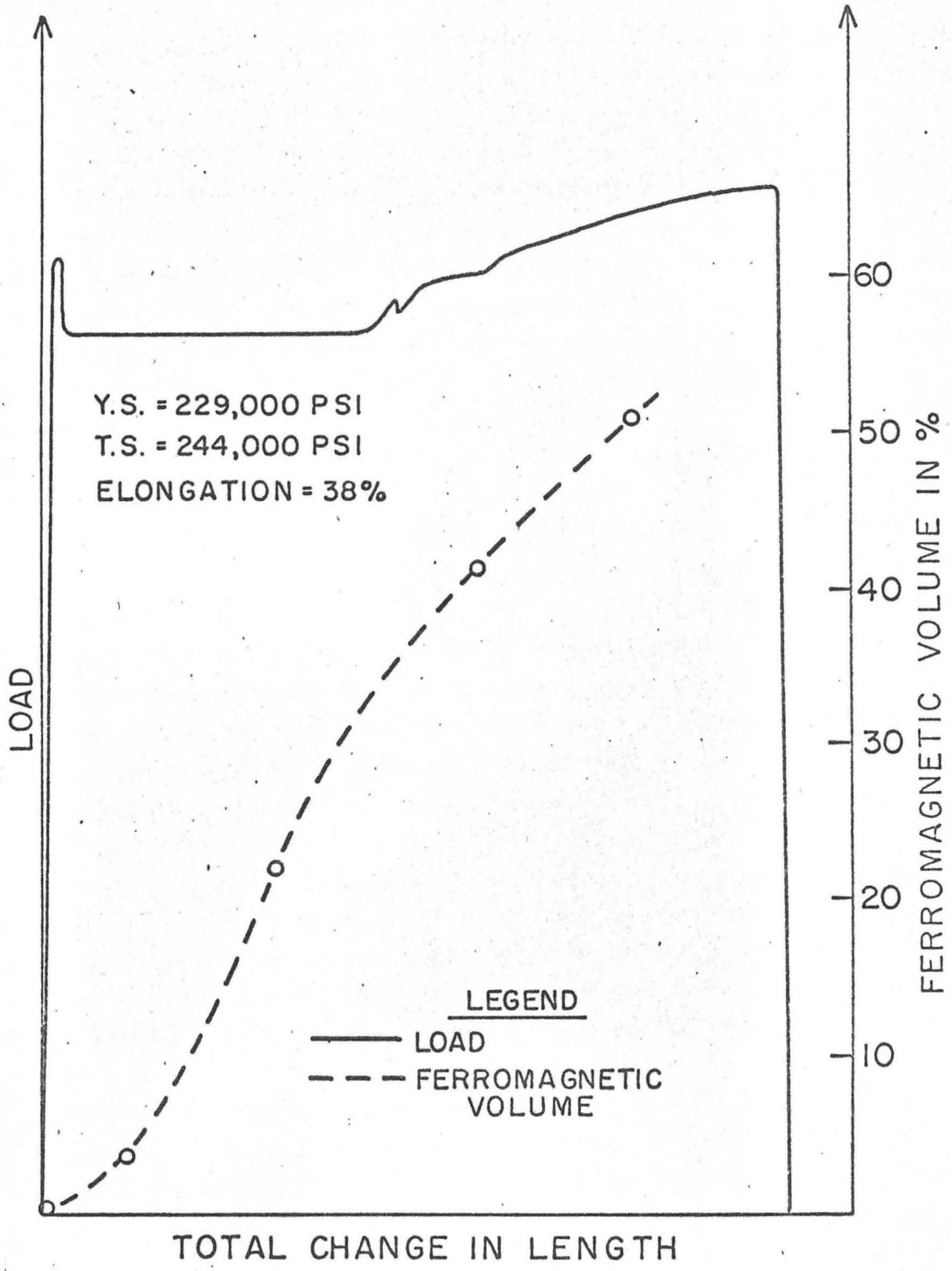
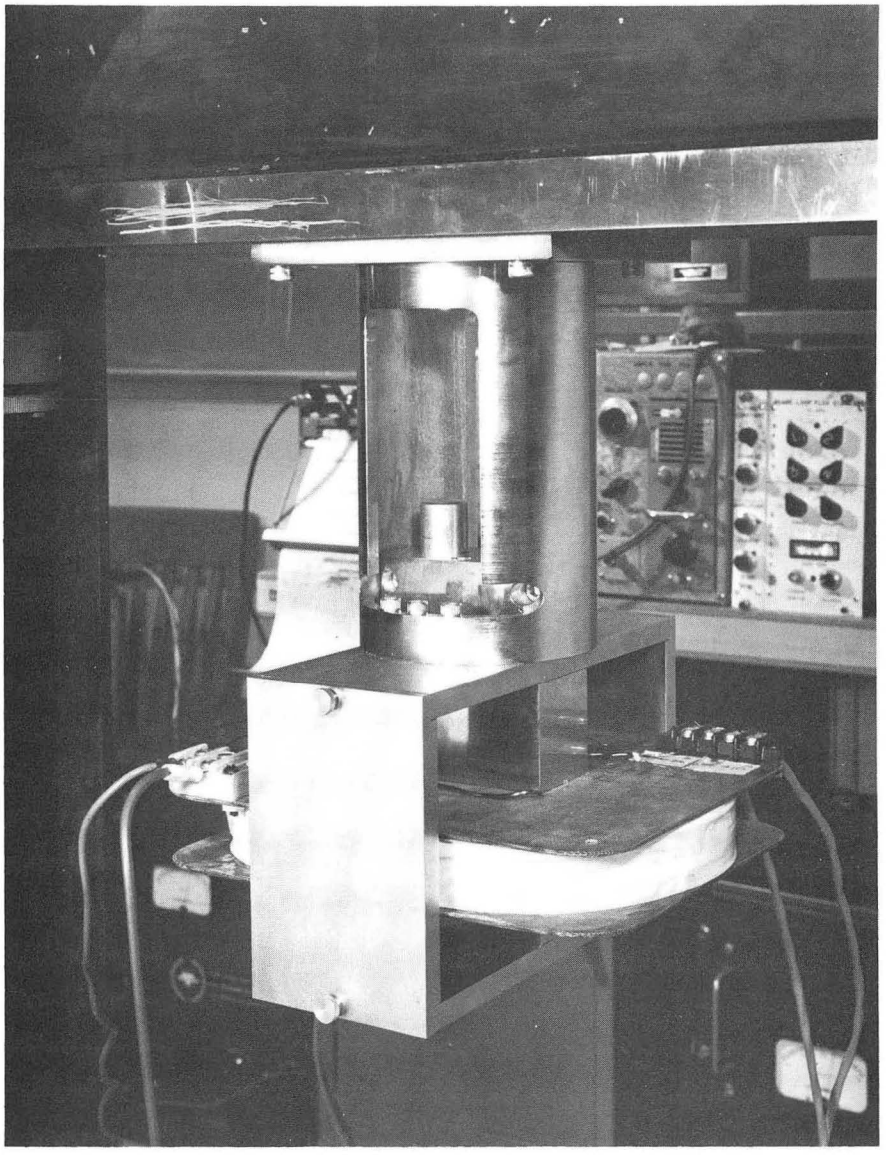


Fig. 27 Alloy II - Load displacement curve and corresponding ferromagnetic volume
Prior treatment 550°C 80%
Short gage specimen



XBB 678-4678

Fig. 28. Permeameter used to measure the saturation induction of the specimens during tensile testing.

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