SSC-103

THE TENSILE YIELD BEHAVIOR OF SHIP STEEL

by

W. S. Owen, B. L. Averbach and Morris Cohen

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September 28, 1956

Dear Sir:

As part of its research program related to the improvement of hull structures of ships, the Ship Structure Committee is sponsoring at the Massachusetts Institute of Technology an investigation of the effect of metallurgical structure on brittle fracture of mild steel. Herewith is a copy of the Second Progress Report of this investigation, Serial No. SSC-103, entitled "The Tensile Yield Behavior of Ship Steel" by W. S. Owen, B. L. Averbach and Morris Cohen.

The project is being conducted under the advisory guidance of the Committee on Ship Steel of the National Academy of Sciences-National Research Council.

Please address any comment concerning this report to the Secretary, Ship Structure Committee.

This report is being distributed to those individuals and agencies associated with and interested in the work of the Ship Structure Committee.

Yours sincerely,

K. K. Cowart

K. K. COWART
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

Serial No. SSC-103

Second Progress Report
of
Project SR-136

to the

SHIP STRUCTURE COMMITTEE

on

THE TENSILE YIELD BEHAVIOR OF SHIP STEEL

by

W. S. Owen, B. L. Averbach and Morris Cohen

Massachusetts Institute of Technology
Cambridge, Massachusetts

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Washington, D. C.
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THE TENSILE YIELD BEHAVIOR OF SHIP STEEL

ABSTRACT

The preyield strain phenomena at room temperature and -195° C are examined in two ship steels, (an ABS class B semikilled steel and a rimming steel designated "project steel E"), as a function of annealing and normalizing temperature and the ferrite grain size. The yield points are measured by an approximately constant strain rate test, the load increasing continuously to the upper yield point. A step-load technique with SR-4 electrical resistance gages is used to determine the elastic limit and the preyield strain. These experiments are supplemented by microscopic observations on prepolished tensile specimens.

The plastic strain prior to gross yielding occupies the "delay time" measured by Clark and Wood and by Krafft. It is shown to be of two types: a microstrain which is established almost immediately when a stress above the elastic limit is applied, and a time-dependent strain (here called "creep"). There appears to be a fairly well defined stress (σ_{creep}) above which creep can be detected in short time tests. At both testing temperatures there is a decrease in the upper and lower yield points, σ_{creep}, and the elastic limit with increasing grain size. This decrease is larger in the E than in the B steel. However, increasing grain size affects the elastic limit less at -195° C than at room

temperature. Only the yield point is a linear function of $d^{-1/2}$, where d is the mean ferrite grain diameter.

At room temperature the microstrain at the upper yield point increases with increasing austenitizing temperature and, in the annealed series, with grain size. This microstrain at -195° C is masked by the extensive creep which occurs. The creep rate increases with testing temperature. At room temperature gross yielding follows the upper yield-point microstrain so quickly that creep is not detected in these experiments. The creep rate and the total preyield strain at -195° C increase markedly with austenitizing temperature for both equiaxed ferrite and Widmanstatten structures, and the grain-size dependence exhibited by the annealed series is a secondary effect. The data are inadequate to detect the effects of chemistry, if any, on the low-temperature micro and creep strains.

Thus, while the stress levels for the initiation of microstrain, creep, and gross yielding are all predominantly chemistry and grain size dependent, the strain rate in each range is primarily a function of the annealing or normalizing temperature employed.

INTRODUCTION

In a uniaxial tension test the yield stress of mild steel depends on the ferrite grain size, temperature, and the rate of loading. The yield stress rises steeply with decreasing temperature[1], but brittle fracture intervenes at some critical temperature[2]. Inasmuch as the brittle fracture stress is not markedly temperature dependent[3], it is not clear whether the yield stress levels off below the critical temperature (implying that cleavage is always preceded by some yielding) or whether cleavage simply supersedes yielding at sufficiently low temperatures.

The value of the critical temperature depends upon the grain size. In fact, the transition from ductile to brittle behavior can be obtained at a given temperature, say -195° C[5], by increasing the grain size. At this temperature brittle fracture of mild steel is always preceded by some plastic strain[6]. Experimental findings are described here which provide information concerning the nature of this strain.

Further, it is well known that gross yielding does not immediately follow the application of the static yield stress[4]. Since the behavior of a moving crack is influenced by the extent and nature of the plastic strain formed ahead of it, the delay time is conceivably an important factor in the crack-propagation problem. The results discussed here lead to a phenomenological description of the events which occur during the delay, both at room temperature and at -195° C. The effects of some metallurgical variables are also outlined.

EXPERIMENTAL METHODS

Preparation of Specimens. Parallel experiments were carried out on two ship steels: one a semikilled steel acceptable under A.B.S. class B and the other, Project Steel E, a rimmed steel with pronounced brittle tendencies. The steels were received in the form of 3/4-in. hot rolled plate and had the following compositions:

Steel	C	Mn	P	S	Si	Cu	Ni	Cr	V	Mo
B	0.16	0.69	0.014	0.028	0.022	0.028	0.015	0.01	0.005	0.01
E	0.22	0.36	0.016	0.031	0.002	0.17	0.13	0.08	0.005	0.025

	Al	Sn	As	H	N	O
B	0.027	0.010	0.001	0.0001	0.0023	0.017
E	0.009	0.012	0.001	0.0001	0.0021	0.006

Standard 0.252-in. diameter tensile specimens were prepared by heat treating oversized blanks. One set of specimens was allowed to cool freely in the air and the other was cooled slowly in the furnace. The austenitizing temperatures were in the range 850--1300° C; at higher temperatures severe pearlite divorcement occurred. All the annealed specimens and those normalized in the lower part of the temperature range had approximately equiaxed ferrite grains. Specimens normalized above 950° C had pronounced Widmanstatten structures. Some of the specimens were banded; this feature was

pronounced after annealing in the temperature range 850--1050° C, being more severe in the B steel. The average ferrite grain diameter in specimens with equiaxial ferrite was measured by a lineal analysis method using a Hurlbut counter[6,7]. A mean ferrite path, a_1, (the average linear intercept between ferrite-ferrite and ferrite-pearlite boundaries) was determined for the Widmanstatten structures. The parameters a_1 and d are related; $d = 1.65\,a_1$.

Tensile Tests at -195° C. The tests were carried out on a 60,000-lb capacity Baldwin Tensile Machine, axiality of loading being insured by loading through long high-carbon steel chains. When electrical resistance strain gages were used, each specimen was fitted with two parallel gages on diametrically opposite sides. These were connected in series to average out small effects due to bending; tests in which the strain from each gage was observed separately revealed a negligible difference. Liquid nitrogen was employed as the refrigerant.

Two types of tensile experiments were carried out at -195° C. In one the load was applied slowly but continuously. In the other it was applied in steps with a measured time interval after each application.

Continuous-Loading Experiments. The specimens were loaded as in the usual engineering tests, the machine being preset so that the strain rate was about 0.005 per minute. At the upper yield point, the load dropped, the test being effectively a constant strain-rate experiment.

The strain over a 2-in. gage length was automatically recorded with a differential transformer gage capable of detecting a change of 0.0001 in. In order to attach this gage securely, it was necessary to make two small indentations in the specimen.

Step-Loading Experiments. In these experiments, designed initially to determine the elastic limit, the load was applied in 50-lb (1000 psi) increments. At each step, the load was held constant for a given time, then released to zero; the residual strain was measured and the load was reapplied. SR-4 type electrical resistance strain gages with a sensitivity of less than 2 microinches per inch were used. The elastic limit was taken to be the stress which first resulted in a positive deviation of the zero-load residual strain. This was in fair agreement with the stress at which a departure from linearity could be detected on the stress-strain curve.

Some runs were carried out to vary the time of application of the load at each step. These clearly demonstrated the necessity of standardizing the loading schedule in any experiments designed to compare the effects of metallurgical variables. In all the tests to be described unless specifically excepted, the load was applied for one minute at each stress level until the elastic limit was detected, and then the time increment was increased to four minutes, the strain at load being measured every thirty seconds.

In addition to the experiments conducted in liquid nitrogen, annealed and normalized specimens of both steels were tested at room temperature with the continuous- and the step-loading technique.

RESULTS

Tensile Tests. In all the continuous-loading tests at -195° C, the upper yield point was clearly defined. All the annealed E specimens broke while the load was dropping from the upper yield point; and consequently, the lower yield point could not be determined. The corresponding B specimens broke either just before or soon after the Lüder's strain was completed. The plastic strain preceding fracture of the annealed specimens was always less than 4%, but the maximum ductility of the annealed E steel specimens was only 0.5%. All the fractures were of the cleavage type. Thus, all of these specimens could be classified as brittle. The normalized specimens were appreciably more ductile, elongations greater than 15% being recorded for E specimens normalized at 950° C. Yet the fractures were of the cleavage type. The fracture stress and ductility of both the annealed and normalized specimens have been described in detail elsewhere[6].

In the step-load experiments at -195° C, the shape of the stress-strain (at load) curve was found to depend upon the time of loading. A comparison between two different schedules is shown in Fig. 1. For one of these, the strain-time curves at successive loads are plotted in Fig. 2. When a standardized loading schedule was adopted, the general features of the stress-strain curves were of the same form for all specimens. On unloading from just above the elastic limit, a microstrain was observed which did not appear to depend on the time of loading (within 30 seconds to 30 minutes). Of course, the microstrain increased with the load itself; but on reaching a value between 5 and 40 microinches per inch, time-dependent strain (creep) was found to set in. The magnitude of the microstrain at the onset of creep (ε_{micro}), the stress at which this occurred (σ_{creep}), and the time-rate of strain were all dependent on the composition and thermal treatment. As an example the stress-plastic strain curves for the E steel annealed at a series of temperatures are shown in Fig. 3. ε_{micro} and σ_{creep} are listed, together with related data, in Table I. At σ_{creep} the strain was a rapidly changing function of the stress; and since relatively large stress increments were used, the values given are approximate.

The creep rate increased as successively higher loads

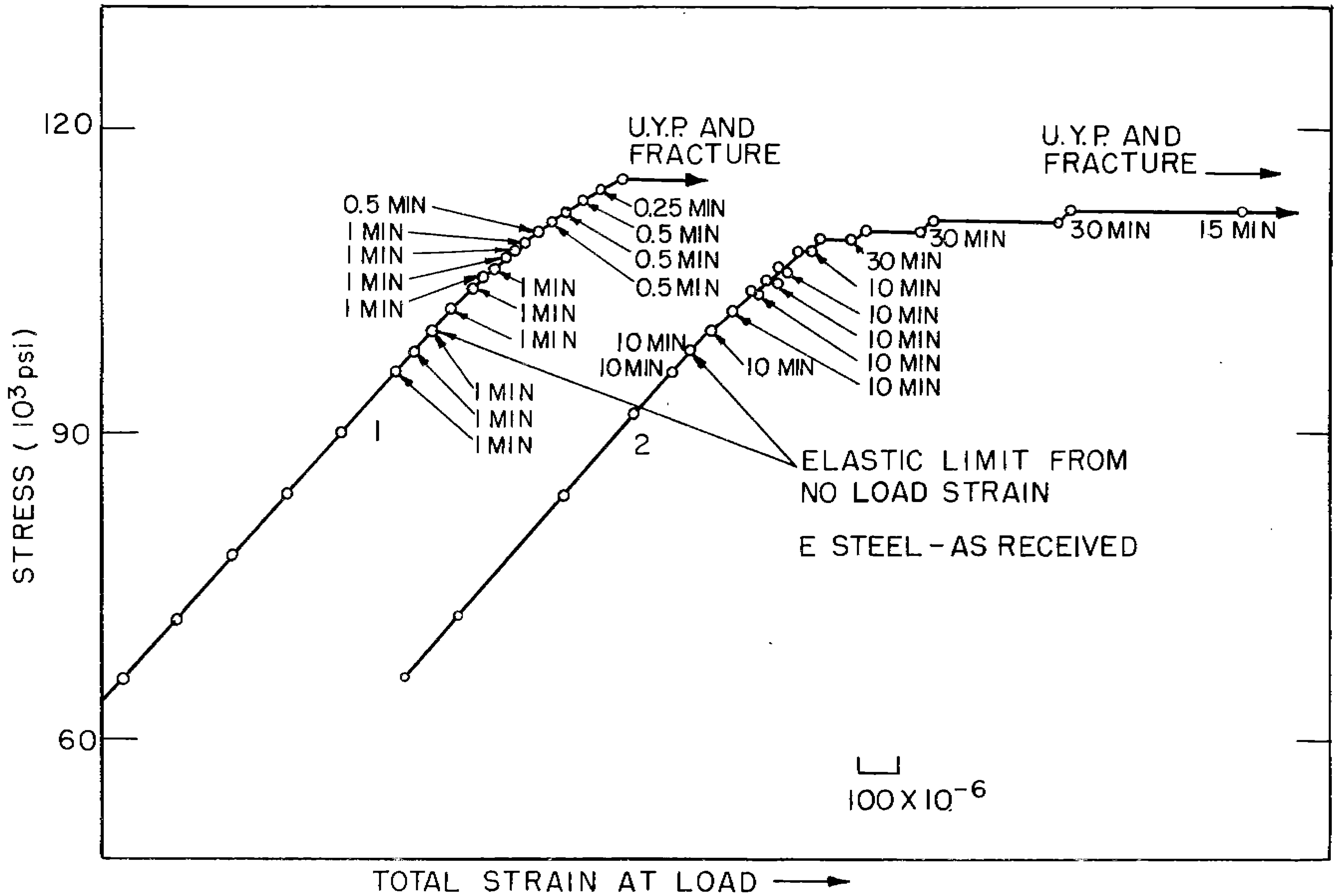

FIGURE I. INFLUENCE OF LOADING TIME ON STRESS-STRAIN CURVE AT —195°C

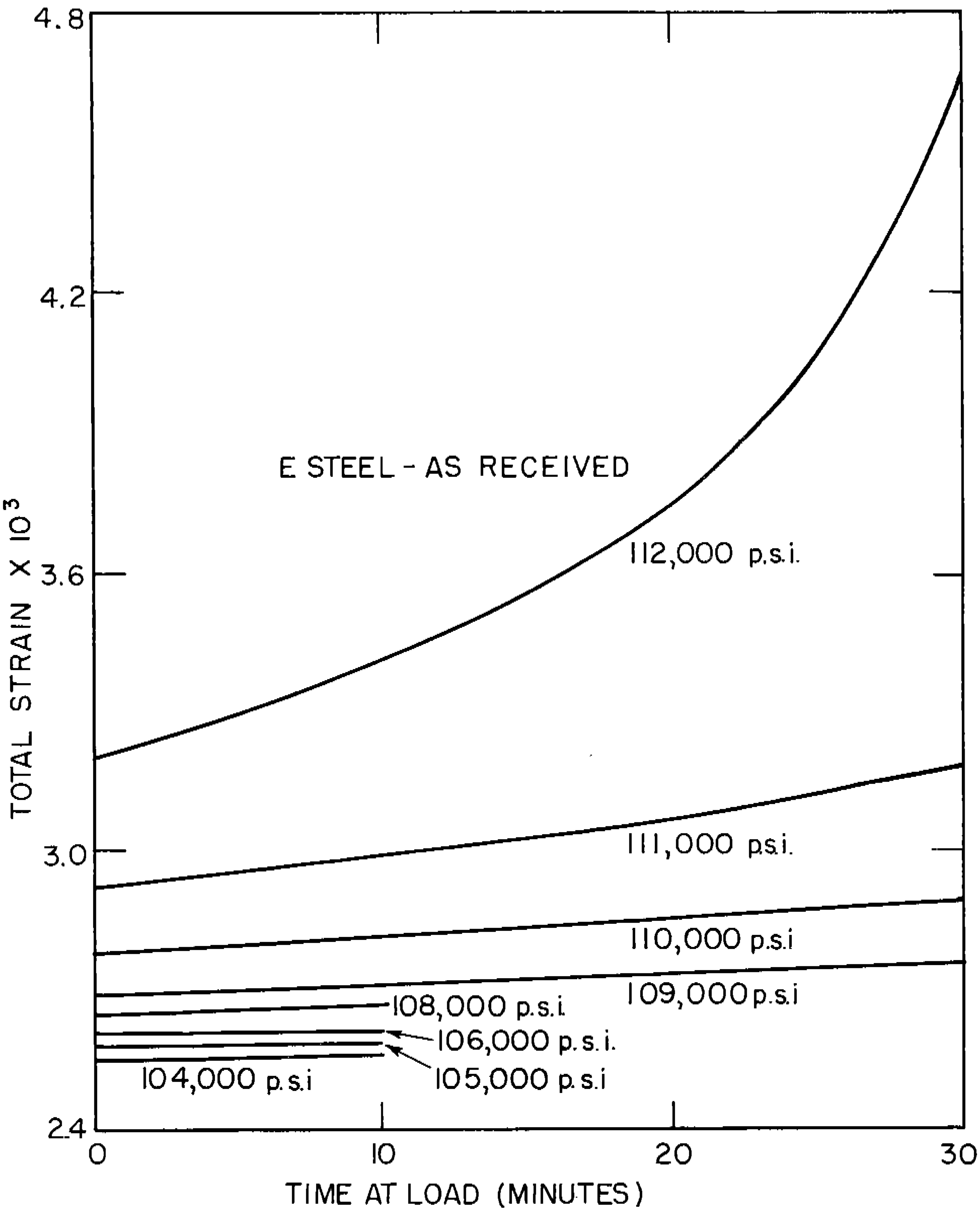

FIGURE 2. CONSTANT LOAD CREEP STRAIN WITH LOAD INCREASED IN STEPS ON THE SAME SPECIMEN–TEST AT −195°C

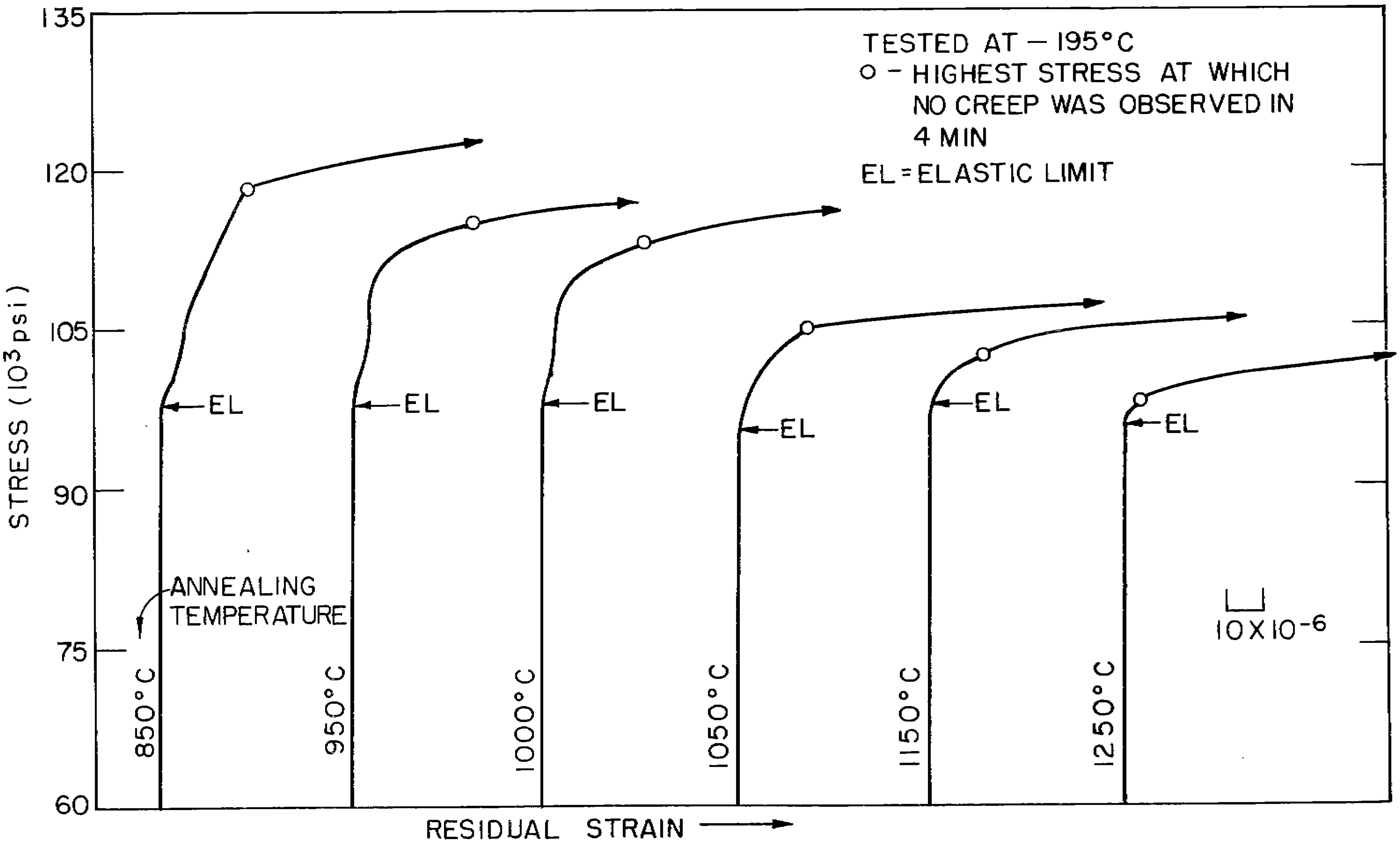

FIGURE 3. STRESS-RESIDUAL STRAIN CURVES FOR E STEEL SPECIMENS ANNEALED AT VARIOUS TEMPERATURES

TABLE I

STRESS AND PLASTIC STRAIN MEASUREMENTS RELATING TO THE ELASTIC LIMIT AND YIELD POINT DEMONSTRATING THE INFLUENCES OF HEAT TREATMENT

	Measured at -195° C					At R. T.
B Steel	σ_{creep} 10^3psi	ε_{micro} (10^{-6})	$\sigma_{creep}-\sigma_{EL}$ 10^3 psi	$\sigma_{uyp}-\sigma_{creep}$ 10^3 psi	ε at σ_{uyp}-2000 psi (10^{-6})	ε at σ_{uyp}-1000 psi (10^{-6})
Annealed						
850° C	103	17	8	11	86	2
950° C	98	12	4	9	286	7
1050° C	92	7	2	11	813	8
1150° C	90	8	4	13	1451	20
1250° C	86	9	4	15	2000	30
Normalized						
850° C	108	10	4	4	35	5
950° C	110	22	12	6	206	7
1050° C	108	31	14	8	414	9
1150° C	108	12	10	10	547	10
1250° C	108	23	10	8	658	12
E Steel						
Annealed						
850° C	118	27	18	14	82	-1
950° C	115	38	15	15	131	-1
1050° C	105	21	9	10	252	-1
1150° C	102	14	4	6	716	8
1250° C	97	3	2	10	1095	10
Normalized						
850° C	124	45	28	8	133	-4
950° C	122	26	20	8	80	4
1050° C	112	29	10	6	311	13
1150° C	108	40	14	8	959	26
1250° C	108	20	12	10	3023	39
B Steel as Received	104	13	8	8	470	15
E Steel as Received	109	32	9	7	759	82

ε = plastic strain measured on unloading.

ε_{micro} = plastic strain at which creep was first observed.

σ_{creep} = stress at which creep was first observed.

were applied until it could not easily be followed on the manually operated strain indicator. The stress at that stage was identified with the upper yield point, being less sharply defined in the large grained specimens. The annealed E steels then strained rapidly to fracture. The other specimens showed an appreciable constant-load strain. The small grained specimens eventually underwent a drop in load to the lower yield point. In some cases this was followed by a small increase in load before fracture.

The preyield strain at the onset of rapid yielding (the upper yield point) appeared to be a function of the metallurgical variables. Unfortunately, it could not be measured precisely in these experiments, but the plastic strain values at a stress 1000 or 2000 psi below the upper yield stress (obtained with the standardized loading schedule) demonstrated the general trends and are shown in Table I. The stress intervals between the elastic limit and the start of preyield creep ($\sigma_{creep} - \sigma_{EL}$) and between this point and the upper yield point ($\sigma_{uyp} - \sigma_{creep}$), shown in the same table, are also approximate due to the relatively large stress increments employed.

No significant difference was detected in the value of the upper yield stress for comparable specimens step-loaded for different times in the range used. However, a marked difference was observed in this stress measured by the continuous- versus step-loading methods. The yield stresses and the elastic limit are shown for the specimens with equiaxed ferrite structure as a function of the mean ferrite grain diameter ($d^{-1/2}$) in Figs. 4 and 5, and for normalized specimens as a function of the austenitizing temperature in Figs. 6 and 7. The values obtained for the elastic limit and lower yield point were found to be reproducible to within ± 2000 psi. The upper yield point data exhibited rather more scatter. All the plotted points are the average of at least two, but usually 3 to 6, determinations.

The room temperature results are given as a function of $d^{-1/2}$ in Fig. 8. Two general features are in sharp contrast to the findings at -195° C[1]. Over a wide range of grain size, the elastic limit was either very close to, or coincided with, the upper yield point. Only with the larger grain sizes (and higher austenitizing temperatures) was there any significant separation between the two properties. Moreover, there was no evidence that preyield creep was a significant factor in these experiments at room temperature.

<u>Metallographic Observations</u>. For the annealed series and specimens normalized between 850° C and 950° C (that is, specimens with equiaxed ferrite structure), the λ_1 parameter (which is directly related to d) varied with heat treatment

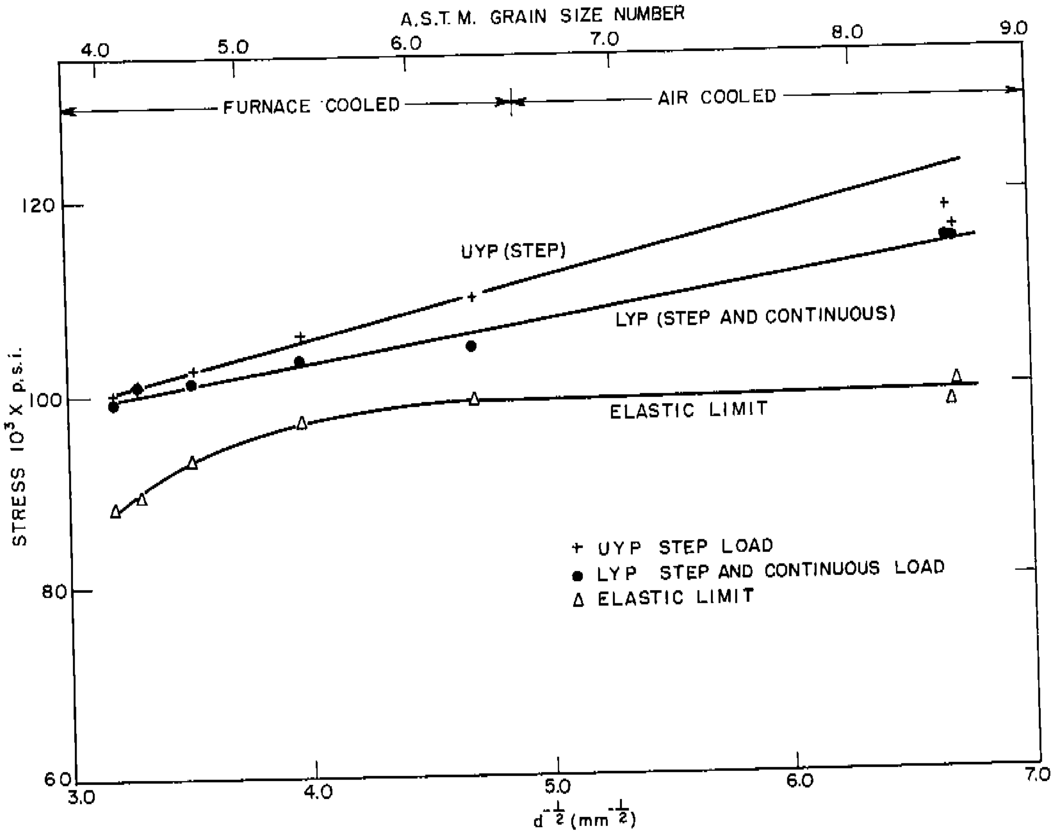

FIGURE 4. YIELD POINTS AND ELASTIC LIMIT AS A FUNCTION OF FERRITE GRAIN SIZE. B STEEL TESTED AT − 195 °C

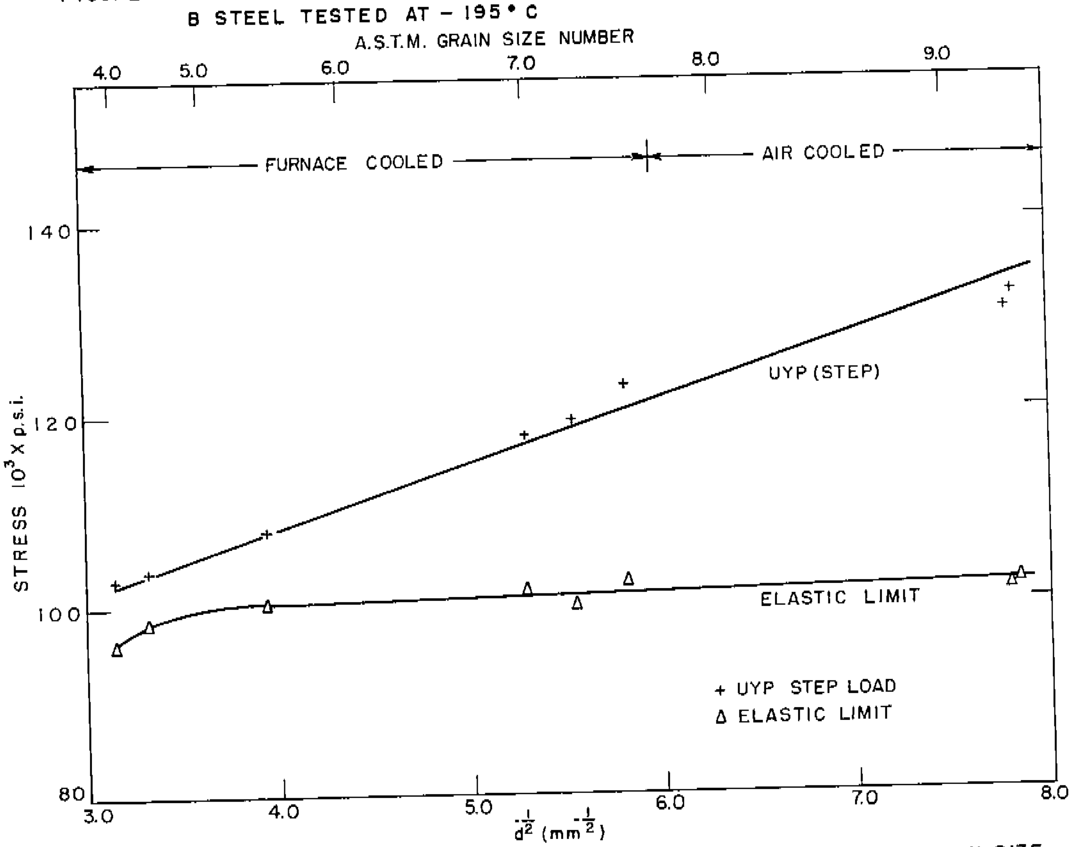

FIGURE 5. YIELD POINT AND ELASTIC LIMIT AS A FUNCTION OF FERRITE GRAIN SIZE. E STEEL TESTED AT − 195 °C

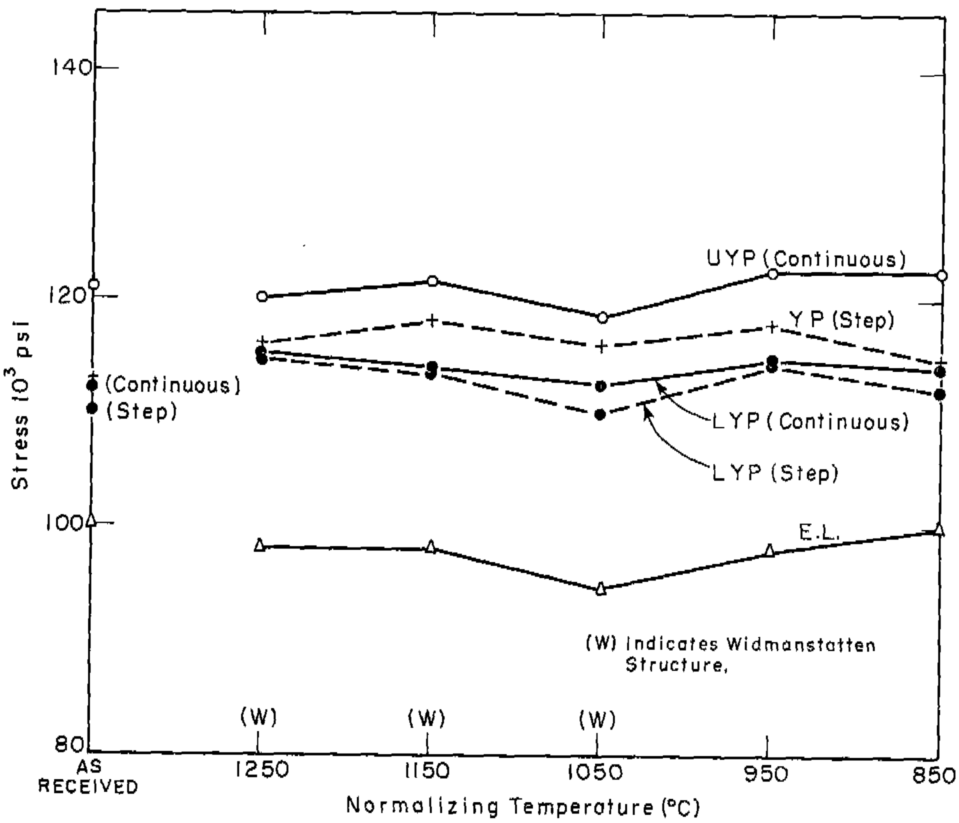

FIG. 6. CONTINUOUS AND STEP LOAD YIELD POINTS AND ELASTIC LIMIT FOR NORMALIZED B STEEL TESTED AT -195°C.

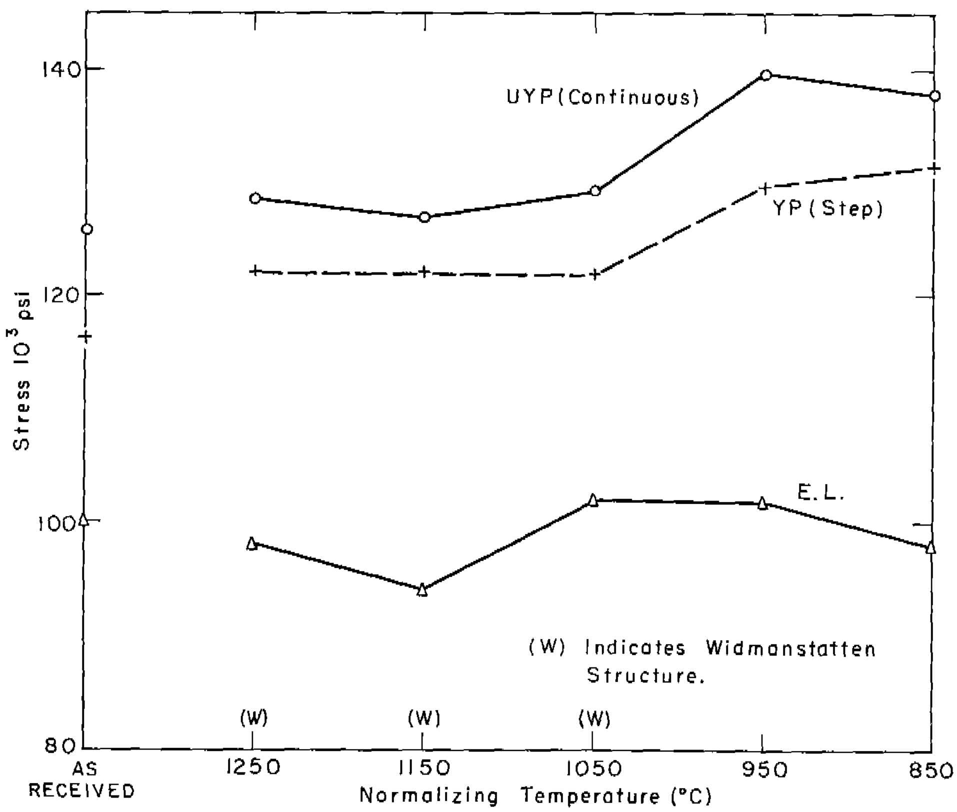

FIG. 7. CONTINUOUS AND STEP LOAD YIELD POINTS AND ELASTIC LIMIT FOR NORMALIZED E STEEL TESTED AT -195°C.

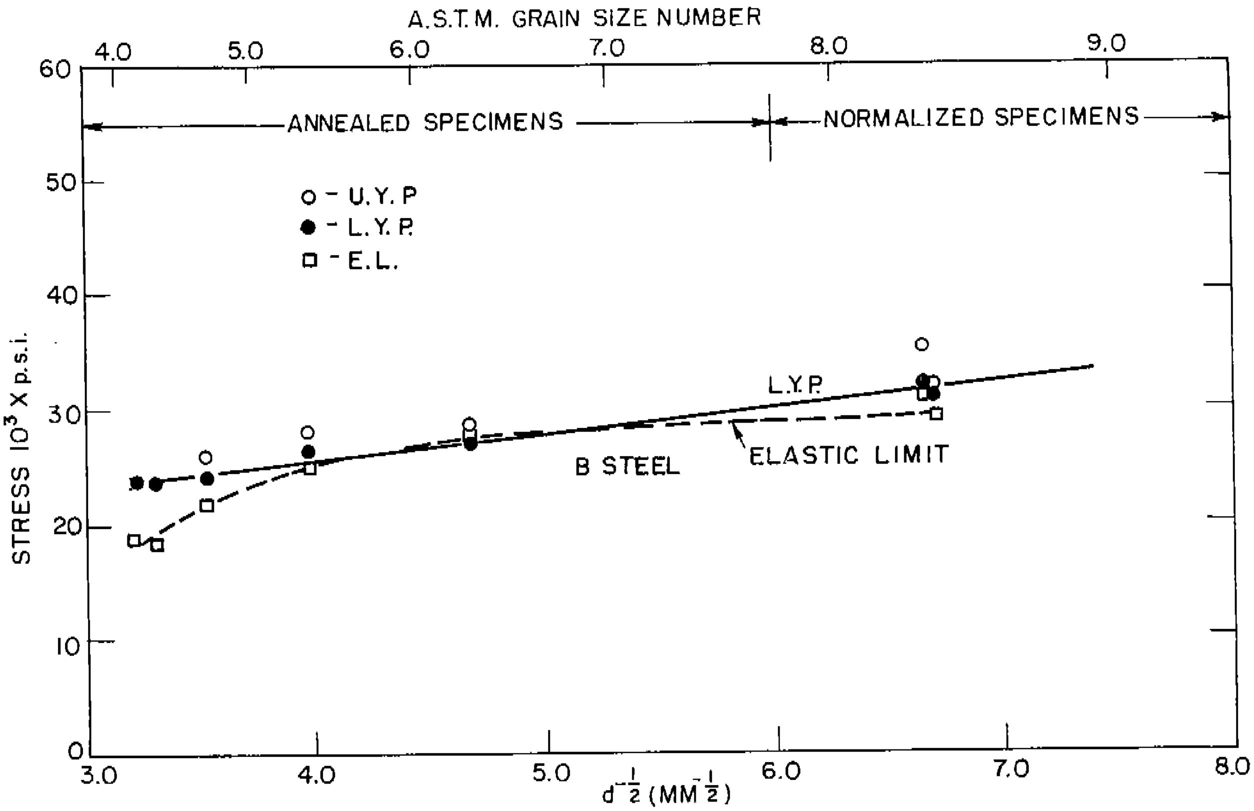

FIGURE 8A. YIELD POINTS AND ELASTIC LIMIT OF B STEEL AS A FUNCTION OF FERRITE GRAIN SIZE-STEEL TESTED AT ROOM TEMPERATURE.

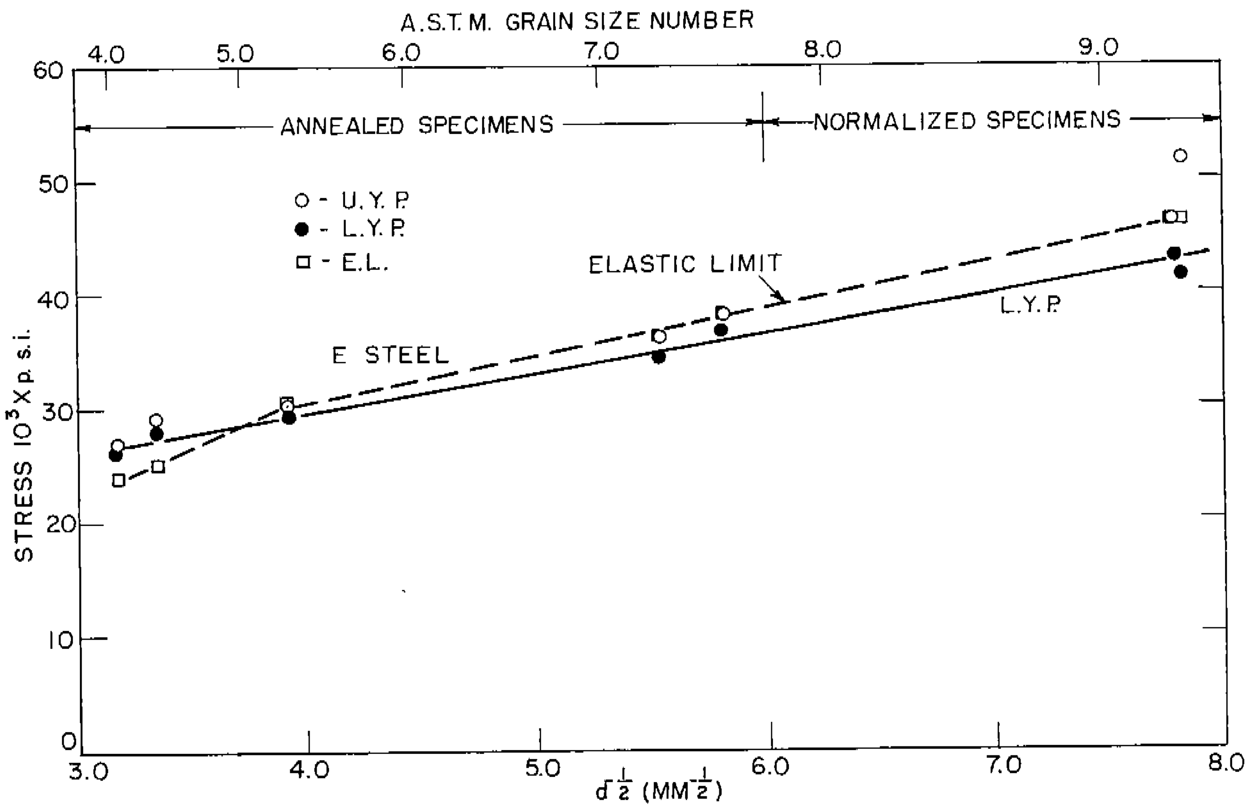

FIGURE 8B. YIELD POINTS AND ELASTIC LIMIT OF E STEEL AS A FUNCTION OF FERRITE GRAIN SIZE-STEELS TESTED AT ROOM TEMPERATURE in the conventional manner. However, it did not change with austenitizing temperature in steels with a Widmanstatten structure.

Metallographic studies were made concerning the nature of the deformation which precedes fracture, two techniques being employed. In one, thin flat tensile specimens were electropolished and etched on one surface, and an electrical resistance strain gage was fixed to the other. The specimens were strained in liquid nitrogen up to a predetermined stress level, then unloaded, and subsequently examined microscopically with stopped-down, oblique or polarized illumination. In other experiments standard round tensile bars were used. Some were loaded to fracture and others unloaded after a measured amount of strain. The tensile bars were sectioned longitudinally and examined by standard methods.

Specimens loaded in the microstrain range (at a stress between the elastic limit and σ_{creep}) exhibited no change under the microscope. Visible slip first appeared after deforming the specimens in the creep range (between σ_{creep} and the yield point). After straining about 200 microinches at a stress 2000 psi below the upper yield point, fine wavy slip was clearly visible in some ferrite grains. On increasing the strain, the concentration of slip lines and the extent of slip increased. After straining beyond the yield point, the surface markings did not appear to be different in nature from those formed by preyield deformation although the area in which slip had occurred was greatly enlarged. Whenever specimens were strained more than about 400 microinches per inch, the usual types of slip markings[6] were observed.

Mechanical twins were present in all specimens after fracture. Usually they were confined to a zone on either side of the fracture surface, the width of the zone increasing with the grain size of the specimen. Only in the coarsest grained samples (those annealed at 1150--1250° C) were twins noted throughout the whole volume of the specimen. Numerous attempts were made to detect the onset of twinning by unloading and sectioning specimens before they had fractured. Preyield twins were found only in the specimens annealed at 1150--1250° C. They were first detected in B steel specimens annealed at 1250° C after straining 200 microinches per inch at a stress 1000 psi below the upper yield point (Fig. 12). Less than 0.2% of the ferrite grains contained twins. The same specimen at the fracture stress contained twins in about 30% of the ferrite grains in areas remote from the fracture. The corresponding E steel gave similar results. In the somewhat smaller grained specimens annealed at 1150° C, no twins were detected until the yield point was reached. In regions remote from the fracture area, the proportion of ferrite

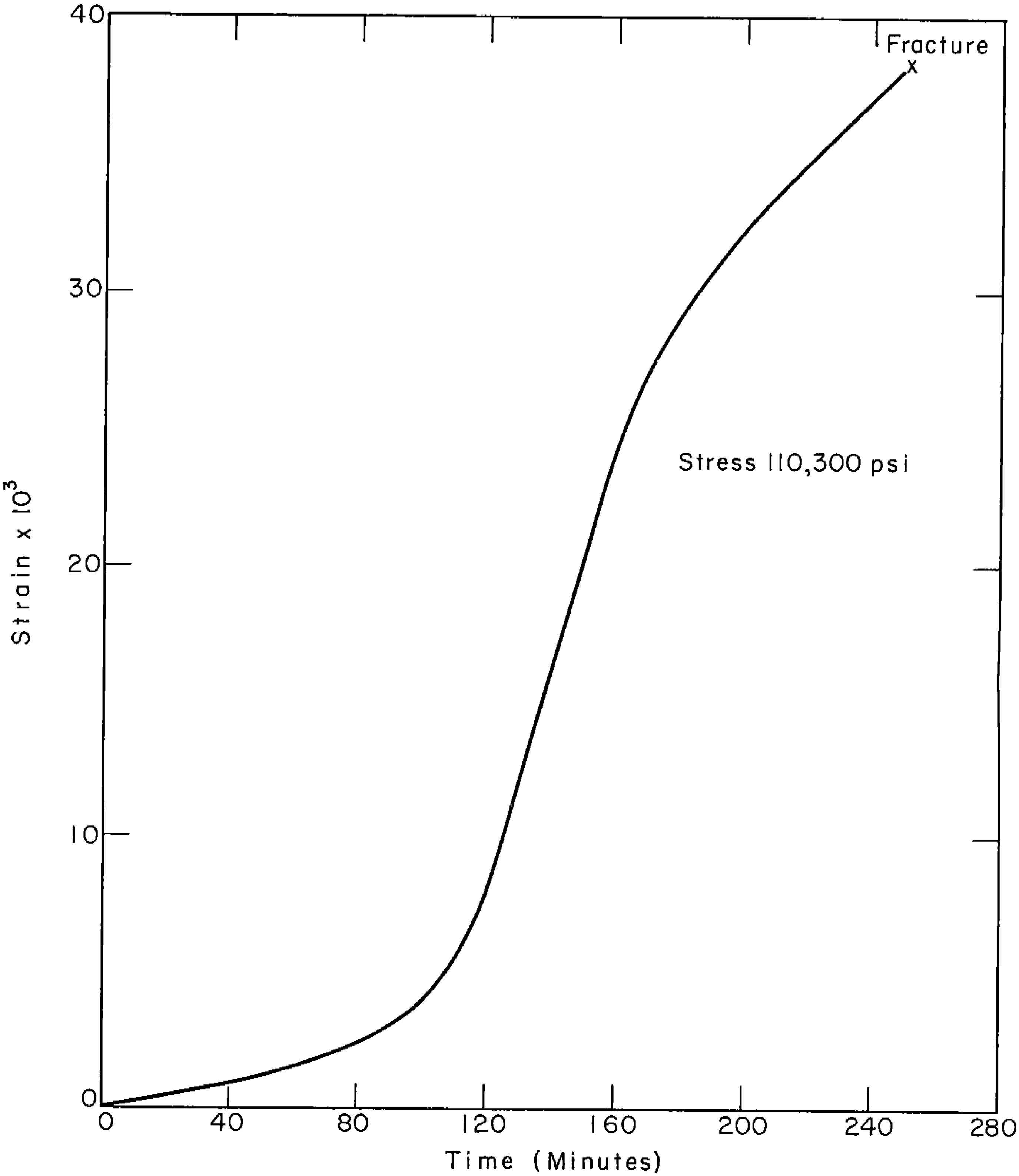

FIG. 9. STRAIN-TIME CURVE AT CONSTANT LOAD FOR E STEEL, AS RECEIVED, TESTED AT −195°C. STRAIN FROM DIAL GAUGE MEASURING CROSS-HEAD MOTION.
(NB. UPPER YIELD STRESS IN STEP LOAD EXPERIMENTS 126,000 psi)

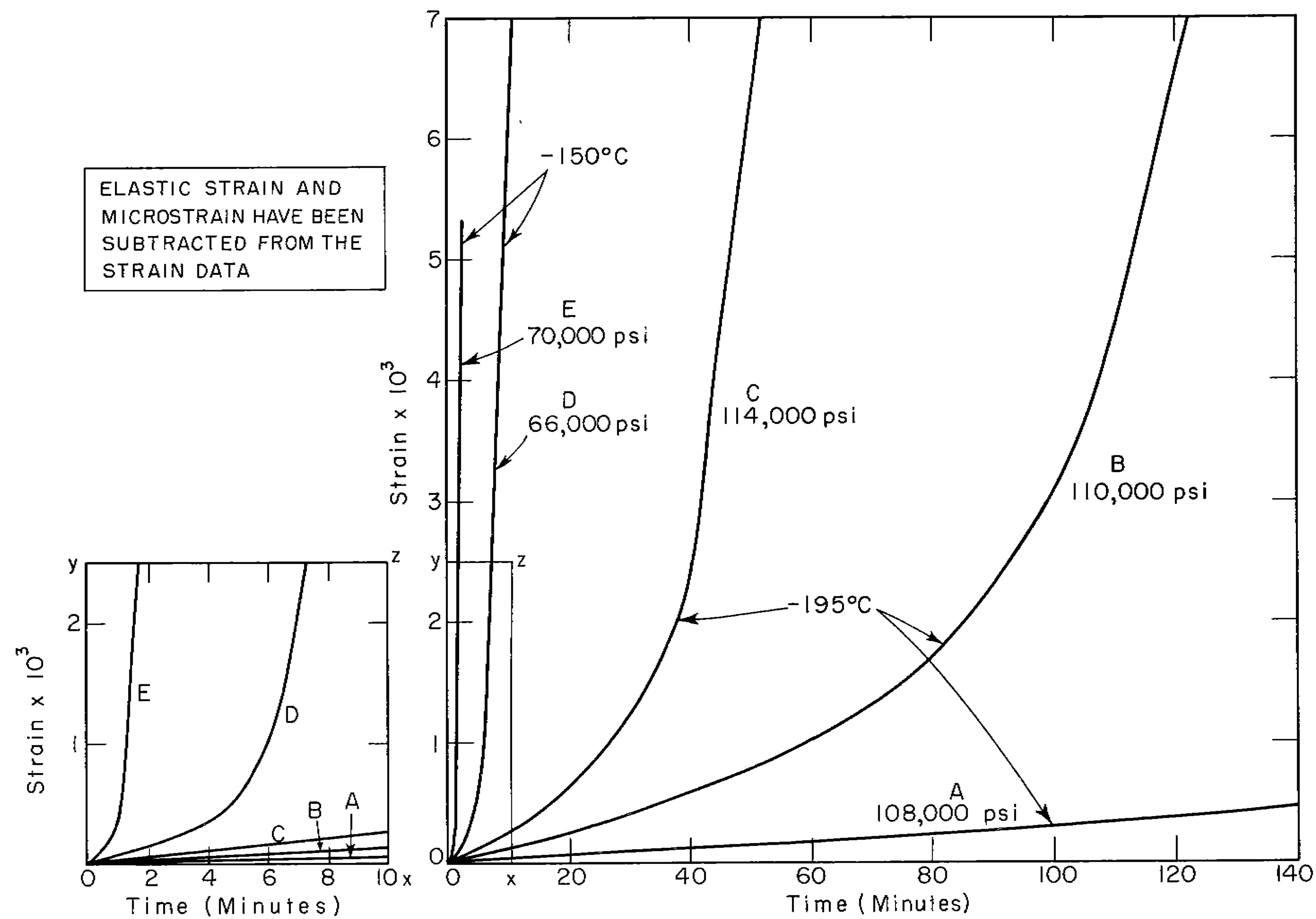

FIG. 10. STRAIN-TIME CURVES FOR CONSTANT LOAD FOR E STEEL, AS RECEIVED, TESTED AT -195°C AND -150°C. STRAIN FROM SR-4 GAUGES ON THE SPECIMEN.

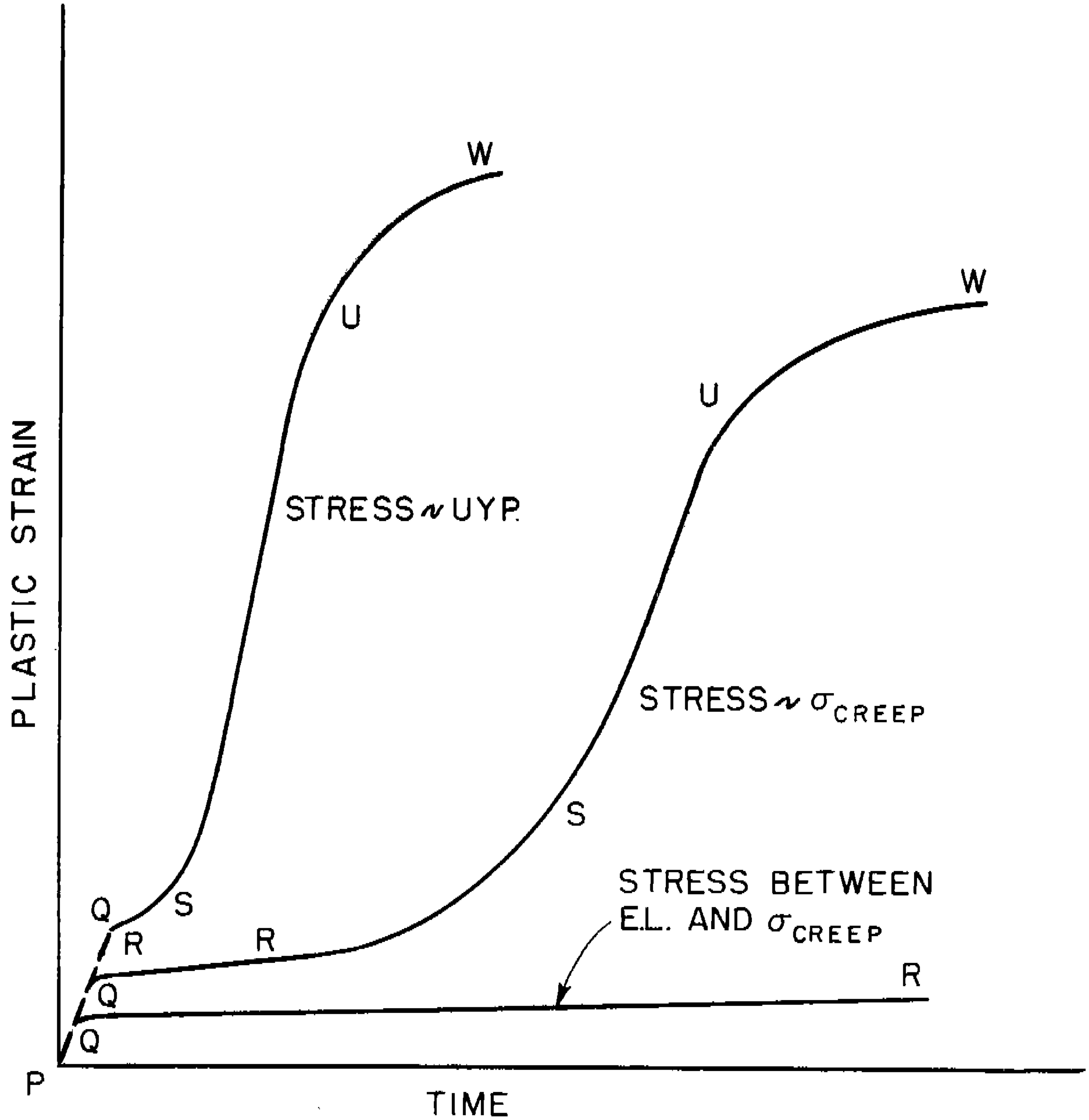

FIGURE 11. GENERAL FORM OF STRAIN-TIME CURVES FOR MILD STEEL TESTED AT CONSTANT TEMPERATURE AND STRESS.

Figure 12. Preyield Twins in E Steel Annealed at 1250°C. Strained 0.0002 at 2000 psi Below Upper Yield Stress in Liquid Nitrogen. X1200

Figure 13. Twins in E Steel after Fracture at -195°C. Surface Electropolished before Testing. X1200

grains containing twins after fracture was appreciably less than in the specimens annealed at 1250° C. With still smaller grained specimens annealed below 1150° C, no twins were found prior to fracture.

Twinning usually was accompanied by visible accommodation slip (Fig. 12). In the vicinity of the fracture, many of the twins had an unusual appearance with one edge exhibiting a saw tooth profile (Fig. 13). This was more evident when electropolished surfaces were examined.

DISCUSSION

At a constant stress and temperature, the onset of gross yielding in mild steel is markedly time dependent(9). A striking demonstration is the threefold increase in the room temperature tensile upper yield stress which can be produced by rapid loading(8). The phenomenon has been studied systematically by Clark and Wood(9--11), who loaded specimens rapidly in tension to a stress greater than the static upper yield stress and, at constant stress, measured a microstrain and an associated time delay before gross yielding. Roberts, Carruthers, and Averbach(12) and Muir, Averbach, and Cohen(13) have demonstrated that a microstrain of about the same magnitude can be measured at a stress a little below the yield point but that a well defined stress, the elastic limit, exists below which no microstrain is detected.

At room temperature the delay time at the static upper yield stress is roughly one second in both tension[10] and compression[21] tests. A rapid loading technique was not used in the M.I.T. work[12,13], and the microstrain appeared to be instantaneous.

Clark and Wood found the delay time for a mild steel at a stress just above the static upper yield stress to be very long at -195° C. The time decreased with increase in stress until a stress was reached above which there was no delay. A similar plateau was found in Krafft's[21] compression tests, and in this case it was identified with the stress at which twinning is the predominant mode of deformation. In the present experiments the maximum stress applied was the static upper yield stress and twinning played only a secondary role. Under these conditions Clark and Wood measured delay times of about 30 minutes (which is ample to explain the elevation of the yield point produced by the continuous-load runs compared to the step-load runs in this investigation). To obtain such long delays, they must have recorded the time required for some unspecified amount of gross yielding. At this temperature, such times are not greatly different from the total time to fracture. In Krafft's experiments no deformation was found at stresses below the twinning plateau because to be detected with his equipment, it was necessary for a strain of at least 150×10^{-6} to occur during the relatively short time the load was applied (about 5 microseconds)[23]. The present experiments indicate that if in either of the previous studies the time for the initiation of a plastic strain of say 10×10^{-6} had been measured the delay times at stresses near the static upper yield stress would have been of the order of seconds. On this basis a phenomenological description of the deformation which precedes gross yielding at low temperatures can now be given.

Three types of plastic strain involved in yielding can be distinguished: microstrain, creep and gross yielding (Lüder's strain). (It is assumed that twinning does not intervene.) Since in the step-load experiments the loading times were limited to 4 minutes, the connection between creep and gross yielding is not immediately obvious. However, the interrelationship was clearly demonstrated by experiments using the as-received E steel in the form of standard 0.252-in. diameter specimens of 2 1/8-in. gage length, with a deadweight machine*. A stress a little below the yield stress as determined by the step-load experiments was applied. The strain was observed as a function of time, by both electrical resistance gages attached to the specimen and a dial gage (capable of detecting 5×10^{-5} in.) measuring cross-head motion.

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*These experiments were performed by R. Asimow at M.I.T. during the summer of 1955.

To illustrate the general shape of the curves, in Fig. 9 the plastic strain (as indicated by the dial gage) at -195° C at a stress 15,000 psi below the step-load yield is shown as a function of time. The microstrain (which could not be measured accurately here) and the elastic strain have been subtracted. The strain rate first increases until, in this case after about 120 minutes, the rate becomes constant and rapid. This corresponds to the generation of the Lüder's strain. After a further period (at 160 minutes, Fig. 9) the strain rate decreases; that is, strain hardening sets in. Soon afterwards, this specimen fractured.

The details of the creep strain are revealed more accurately by the SR-4 electrical strain gage readings. The data for three different loads at -195° C and two at -150° C are plotted in Fig. 10. Again, the elastic strain and microstrain have been subtracted.

These experiments show that:

1. The creep strain will, if sufficient time is allowed, accelerate to gross yielding.

2. There is no discontinuity between the measured creep and Lüder's strains.

3. The actual static upper yield stress at -195° C is even lower than that indicated by the step-load experiments.

4. The creep phenomenon can be detected at -150° C as well as at -195° C.

The sequence of events leading to gross yielding can be discussed by referring to a generalized constant stress strain-time curve (Fig. 11). As soon as the load is applied, the strain increases from P to Q and remains constant for some time. The microstrain PQ is smaller, the lower the applied stress, the vanishing point being the elastic limit. The accuracy with which this point is measured depends upon the sensitivity of the strain detection; but since the microstrain persists at stresses near the elastic limit, it does not depend upon the loading time within a wide range.

The true elastic limit might be defined as the stress necessary to activate the first dislocation mill (e.g., Frank-Read source). It is usually considered that a source is located within the lattice, and hence this stress should be insensitive to grain size. To activate the source, a critical stress is required, which is assisted by thermal fluctuations[17,18]. In the past the required applied stress has been equated to the upper yield stress. This is no longer tenable because the yield stress is grain-size dependent (Figs. 4 and 5). It seems more reasonable to associate the activating stress with the elastic limit, implying that the true elastic limit is unaffected by grain size. However, when measured at room temperature with a strain sensitivity of about 2×10^{-6}, it is markedly dependent upon grain size (Fig. 8), as well as upon steel chemistry (although changing from furnace to air cooling produces no effect). At -195° C, for a variety of heat treatments the measured elastic limits of the two steels lie within a fairly small stress range (Figs. 4, 5, 6 and 7), but systematic changes with grain size can be detected (Fig. 4). This dilemma has not been resolved, and the effect of metallurgical variables on the elastic limit requires further investigation.

Returning to the constant-stress curve in Fig. 11, after some time, R, a detectable increase in strain occurs. The strain rate then increases in the creep range until at time S a steady state is reached, corresponding to the generation of the Lüder's strain, and the rate of increase of strain is rapid and constant. At a stress a little above the elastic limit, the time required to reach the point R is greater than it was practical to employ in the present experiments. Consequently, it has not been determined whether the creep can be initiated at all stresses above the elastic limit if sufficient time is allowed or whether a critical stress is required. However, experimentally there is a fairly well defined stress (σ_{creep}) at which the onset of creep (point R) can first be detected in a relatively short time (say, 4 minutes). At -195° C, σ_{creep} decreases with increasing ferrite grain size, but the normalizing temperature in the range 1050--1250° C (giving Widmanstatten structures with roughly the same a_1 value) has little effect (Table I). For comparable heat treatments, σ_{creep} is higher in the E steel than in the B.

At stresses above σ_{creep}, the onset of creep occurs so soon after the load is applied (QR becomes very small) that the microstrain cannot be separated from the creep strain. However, it can be estimated by extrapolating the microstrain measured between the elastic limit and σ_{creep} (Fig. 3) to higher stress levels. The microstrain at the upper yield stress appears to be larger at -195° C than at room temperature presumably because the microstrain at -195° C occurs in a higher stress range. At room temperature the microstrain at the upper yield stress increases with austenitizing temperature, and in the annealed series, with grain size. The low-temperature extrapolated data are too crude to reveal any trends.

The time interval RS (Fig. 11) decreases with increasing stress. At the upper yield point gross yielding starts very soon after applying the load. In the step-load experiments at -195° C if the point S was reached within 4 minutes, the stress was recorded as the upper yield stress. The stress range σ_{uyp}--σ_{creep} appears to be almost the same for the two steels regardless of the austenitizing temperature but is slightly larger for the annealed than the normalized steels. Consequently, the time in this stress range is roughly the same in each test. However, the strain at S increases markedly with austenitizing temperature. (The strain, including a very small microstrain, at 2,000 psi below the upper yield stress is listed in Table I.) Thus, the creep rate must also increase with austenitizing temperature. The trend is the same in both the annealed and normalized series, and so it is deduced that the effect is dependent on austenitizing temperature rather than on grain size. This suggests that submicroscopic effects have an important bearing on the creep process. Furthermore, for the annealed series the creep occurs in a lower stress range, the higher the austenitizing temperature; thus the effect is not the result of differences in applied stress. The experimental data are inadequate to distinguish between the behavior of the B and E steels in this connection.

The creep strain (Fig. 10) and the stress range in which it can be observed both decrease with increasing temperature. At the upper yield stress at room temperature, only a very small microstrain (even less than 2×10^{-6} for small grain size specimens) and no creep were detected in the present experiments. The step- and continuous-load yield points were the same. Thus, it seems the points Q, R and S almost coincide in this case. However, when high speed loading and precise time measurement are employed, a creep range can be detected[23]. Evidence of its existence can be found in data published by Clark and Wood and by Krafft.

If a stiff machine with constant strain rate is employed, the onset of gross yielding (S) is accompanied by a drop in load to the lower yield point. The hydraulic machine used in the step-load experiments had a relatively sluggish load response; nevertheless a drop was always observed soon after gross yielding started. It may be that, with a very stiff machine and a rapid response load-indicating system, a drop could have been detected even during the creep stage.

Both the upper and lower yield points are affected by chemistry and increase with decreasing grain size but do not depend on whether the structure is established by annealing or normalizing provided Widmanstatten structures are avoided. The stress for creep initiation and the elastic limit have similar characteristics. However, unlike the elastic limit, the upper and lower yield stresses are linear functions of $d^{-1/2}$ (Figs. 4, 5 and 8). The data also suggest that this relationship does not hold for the creep-initiation stress but are not sufficiently accurate to decide this point with certainty.

Hall, Sylvestrowicz[15] and Hart[22] have shown that the room temperature Lüder's strain is generated with constant velocity at the lower yield stress. It is found here that the Lüder's strain rate is also constant at low temperatures when generated at constant load. When appreciable strain hardening sets in (UW in Fig. 11), the strain rate decreases if the applied load is constant; or if a constant strain rate is used, the load increases.

Paxton[19,20] has reported that, in single crystals, the total Lüder's strain (SU) increases as the testing temperature is lowered. Static load-strain curves for mild steel published by Clark and Wood[10] show that the Lüder's strain increases from 2% to 3% on reducing the temperature from 23° C to -60° C. In the step-load experiments at -195° C, the total Lüder's strain frequently could not be measured because fracture intervened. However, in the more ductile specimens strains greater than 5% were observed before any detectable increase in load occurred. No systematic study of this effect appears to have been made in polycrystalline steel.

Experiments of the type discussed above give no information about the micromechanisms involved in the various stages of yielding. The metallographic study of small prepolished specimens shows that twinning is not operative in room temperature yielding. However, at -195° C, twinning and slip occur together in the ship steel specimens with large grain size. The appearance of twins does not coincide with the elastic limit; evidently, twinning is preceded by some slip. Moreover, the contribution of twinning to the total strain does not appear to be a major factor. No twins were found in the specimens with small grain size even after gross yielding (but not fracture) had occurred. In both steels when annealed above 1050° C, twins were detected during the creep stage. It may be significant that these are the heat treatments which result in rapid and extensive creep. However, very few twins are formed before gross yielding, and it seems improbable that these can account for more than a small fraction of the creep strain.

The slip markings on prepolished surfaces are of the same type whether they are formed in the creep or Lüder's strain range, nor are they visibly affected by changing the temperature from room temperature to -195° C. Since the measured strain was an average over at least a 1-in. gage length and creep and Lüder's strain could not be distinguished from each other by metallographic methods, the distribution of strain on a macroscale during the yielding sequence was not revealed by the present experiments. It is not known whether creep and Lüder's strain are continuous or competitive processes. The first appearance of creep may coincide with the start of a complex Lüder's pattern which then spreads through the specimen with increasing velocity until a steady state corresponding to the Lüder's strain is reached. Alternatively, a small strain, increasing with time, may first be produced throughout the specimen establishing conditions which permit the Lüder's strain, subsequently initiated at the shoulders, to advance and sweep up the preyield strain. Attempts to differentiate between the two mechanisms are now being made by using long flat specimens in which a simple Lüder's band is generated from each end.

CONCLUSIONS

During the time interval between the application of a stress in the vicinity of the static yield point and the onset of gross yielding of mild steel, that is, the delay time as measured by Clark and Wood and by Krafft, two preliminary stages of plastic strain take place. If the applied stress is greater than the elastic limit, a microstrain (always less than about 50×10^{-6}) appears almost instantaneously. This is followed by the slow generation of strain, here called "creep", which may reach an appreciable value before gross yielding occurs. Previously, although microstrain has been associated with the delay time, the importance of the creep period does not seem to have been recognized.

Experiments with two ship steels (ABS class B and project steel E) tested at room temperature and -195° C reveal:

1. The elastic limit, like the yield stress, depends upon the steel chemistry, increases with decreasing temperature and grain size, and is not affected by changing from furnace to air cooling. However, unlike the yield stress, it is not a linear

function of $d^{-1/2}$, and the chemistry and grain size effects are greatly reduced at -195° C.

2. At room temperature, the microstrain at the yield point increases with austenitizing temperature in both the annealed and normalized series and is smaller than at -195° C. There is little difference between the two steels in this respect.

3. At low temperatures creep can be detected over an appreciable stress range, and at -195° C can occupy a period of 30 minutes or more. However, at room temperature the creep stage is obscured by the early initiation of gross yielding.

4. At -195° C there is a minimum stress at which creep can be observed in a relatively short time (say, 4 minutes). This stress shows the same trends as the yield stress and elastic limit. It is affected by steel chemistry and is predominantly dependent on grain size.

5. The creep rate and total creep strain at -195° C are markedly increased by increasing the austenitizing temperature in both the annealed and normalized series. As with the room-temperature microstrain at the yield point, these effects are not primarily grain-size dependent.

6. Twinning at -195° C is preceded by some slip, and in general it is not a significant factor in preyield deformation. However, in coarse-grained steels at -195° C twins are formed during creep, which in this case is rapid and extensive. The slip

occurring during creep cannot be distinguished microscopically from that resulting in gross yielding.

At room temperature the E steel has a higher elastic limit and yield strength than the B steel with the same grain size or same austenitizing treatment. The differences in yield strength persist at -195° C, and the stress at which creep is detected is also greater for the E steel. The elastic limits approach a common value at -195° C. However, at both temperatures the total microstrains and preyield strains show no appreciable differences between the two steels.

The heat treatments appear to influence the critical stresses (i.e., the elastic limit, σ_{creep} and the upper and lower yield points) and the total strains (i.e., the microstrain at the upper yield stress and total creep strain) by fundamentally different mechanisms. The stresses are primarily affected by ferrite grain size; but the strains appear to be determined by submicroscopic conditions determined, at least in part, by the austenitizing temperature, with grain size correlations being secondary.

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