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SR-136

BRITTLE FRACTURE OF MILD STEEL IN TENSION AT -196 C

by

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

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November 5, 1957

Dear Sir:

Under its research program directed toward improving ship steels, the Ship Structure Committee is sponsoring at Massachusetts Institute of Technology a fundamental investigation of the influence of metallurgical structure on properties of steel. One phase of this work has been an examination of low-temperature behavior. Herewith is the Third Progress Report, SSC-109, of this project, entitled "Brittle Fracture of Mild Steel in Tension at -196 C," by W. S. Owen, B. L. Averbach, and M. Cohen.

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

This report is being distributed to individuals and groups associated with or interested in the work of the Ship Structure Committee. Please submit any comments that you may have to the Secretary, Ship Structure Committee.

Yours sincerely,

KKCowart

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

Serial No. SSC-109

Third Progress Report
of
Project SR-136

to the

SHIP STRUCTURE COMMITTEE

on

BRITTLE FRACTURE OF MILD STEEL IN TENSION AT -196 C

by

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

Massachusetts Institute of Technology
Cambridge, Massachusetts

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Washington, D. C.
National Academy of Sciences-National Research Council
November 5, 1957

BRITTLE FRACTURE OF MILD STEEL IN TENSION AT -196 C

ABSTRACT

The tensile fracture behavior of a mild steel at -196 C has been studied in some detail. With the aid of long thin strip specimens loaded at controlled crosshead speeds between 8.9×10^{-4} and 1.6×10^{-1} in./min, the strain pattern and microscopic changes preceding fracture were observed, and the magnitude local strain was measured. Specimens heat-treated to alter the tendency toward brittle behavior, but maintaining ferrite-pearlite structures, were also examined.

Under these conditions, all the Lüder's bands display microcracks in some ferrite grains. However, the as-received and normalized specimens do not fracture at low yield stresses (slow loading speeds) during the spread of the Lüder's bands. By raising the loading speed, a critical stress is reached when fracture occurs after a delay time. The formation of microcracks and fracture is always preceded by gross yielding. On further increasing the loading rate, the fracture stress rises along with the yield stress.

Deductions from the dislocation pile-up theory of fracture in polycrystalline metals are not compatible with the experimental data. It is concluded that the microcrack model suggested by Low is more appropriate.

Some observations on the creep occurring in Lüder's bands during their propagation at -196 C are included.

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INTRODUCTION

Most recent discussions of brittle fracture in polycrystalline metals[1--7] have utilized a model elaborated by Stroh[1, 2] and applied by Petch[3, 4], in which it is assumed that the local stress concentration necessary to cleave a crystal is produced by an extensive defect in the form of dislocations piled up against a grain boundary. It has been implied[2, 7] that this is, in effect, a restatement of the Griffith[8] concept of fracture in terms of dislocation theory. However, Low[9, 10] maintains that, at least in mild steel, real cracks of sub-critical size are actually formed before the main fracture occurs and, thus, a more direct application of the Griffith model (as modified by Orowan[11]) is appropriate

The experimental evidence supporting these theories is not extensive. Stroh and Petch rely principally upon two related experimental observations: that the fracture stress (σ_F) at -196°C is linearly related to the reciprocal of the square root of the mean grain diameter (d), and that the proportionality constant so determined is of the expected magnitude[2]. However, Low[9] demonstrated the existence of subcritical cracks in iron deformed by slow bending at -196°C; he also found a linear σ_F vs. $d^{-1/2}$ relationship by tension tests and deduced a value of the proportionality constant which is compatible with the Griffith-Orowan theory. Clearly, comparisons over a wider range of experimental data are required to assess the applicability of these theories to the low-temperature fracture of mild steel. In the present work, the search for subcritical microcracks is extended to specimens tested in tension, and the predictions of the two theories are compared with the experimental fracture stress and strain as a function of the loading rate.

As the stress is increased slowly in a simple tensile test at -196 C, four stages of nonelastic strain are identifiable. First, there is a preyield microstrain which is less than about 50×10^{-6}. This is followed by a microcreep strain, occupying an appreciable time interval before it is replaced by a third stage--the generation of Lüder's bands[12]. The -196 C microcreep strain may be as large as 10^{-3}, and is accompanied by the appearance of fine slip markings. During the propagation of the Lüder's bands, the stress is reasonably constant, but it increases again in the fourth stage (strain hardening) after bands have covered the whole specimen. Subsequent fracture is still predominantly by cleavage, and it was in this range that Low[10], using a slow-bend technique, observed microcracks which did not propagate until the stress was further increased. Petch[4] also obtained much experimental data in this range. Considering that such specimens exhibit appreciable ductility before final rupture, it would seem more appropriate to examine the theories of brittle fracture in relation to data for which the plastic strain preceding fracture is a minimum.

Mild steel specimens that fracture at -196 C with very small overall ductility are readily obtained by suitable heat treatment, but relatively few accurate measurements of the nonelastic strain have been reported. During the spread of a Lüder's band, the deformation is heterogeneous on a macroscale as well as on a microscale, and observations over a 1- or 2-in. gage length give no indication of the strain in the vicinity of the fracture. Furthermore, the recorded strain is dependent upon the geometry of the specimen. In the present experiments, a technique was employed by which the local strain pattern could be observed both in the vicinity of the fracture and in locations of similar geometry where no fracture occurred.

EXPERIMENTAL METHODS

The mild steel* investigated here was of the rimmed type and was received in the form of a hot-rolled 3/4-in. plate. The effects of a wide variety of heat treatments on this material have been studied in detail using Charpy V-notch and low-temperature standard tensile tests[13]. It has been found that the -196 C properties can be varied from extreme brittleness (950 C furnace cooled) to excellent ductility (950 C air cooled), all with normal equiaxed ferrite plus pearlite microstructures. Specimens treated in this way and some in the as-received condition were used in the present series

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*Analysis.
C 0.22, Mn 0.36, P 0.016, S 0.031, Si 0.002, Cu 0.17, Ni 0.13, Cr 0.08, V 0.005, Mo 0.025, Al 0.009, Sn 0.012, As 0.001, O 0.006.

of tests. The standard-test data are shown in Table I.

The heat treatments were carried out in a helium atmosphere on full-plate-thickness blanks. The specimens, flat strips 0.035--0.040 in. thick with a parallel-sided gage section 5 in. long and 0.25 in. wide, were machined from the center. The last 0.010 in. were removed by slow grinding with a well lubricated porous wheel. Deep etching revealed no evidence of overheating. The specimen and plate surfaces were mutually parallel, so that the specimen axis coincided with the rolling direction. One surface of each specimen had a ground finish with all the scratches (corresponding to about 0 emery paper) running longitudinally. The other was taken down to 000 paper before being polished with diamond paste. An accurate scribing machine was used to mark a scale with 1-mm intervals on the ground side along the whole length of the specimen, the short transverse scratches being hardly deeper than the longitudinal grinding marks.

The tests were conducted with a modified Hounsfield tensile machine. The load was applied through an electrically driven screw and gear train and measured by the deflection of a stiff spring. The overall rate of strain was the independent variable. The range of crosshead speeds lay between 8.9×10^{-4} and 1.6×10^{-1} in./min. The resulting yield stress was indicated on a semi-automatic load-extension recorder.

If the specimen did not break, the test was discontinued when the Lüder's bands covered almost the whole gage section (yielding time 40--60 min). After each run the specimen was removed from the liquid nitrogen bath,

TABLE I

Standard Test Data for Mild Steel*

Heat treatment	Charpy V-notch Transition temp. °C	Yield stress at -196°C Round-bar Tensile test, 10^3 psi	Fracture stress at -196°C Round-bar Tensile test, 10^3 psi	Reduction in area at -196°C Round-bar Tensile test, %	Ferrite grain size A.S.T.M. No.
950° C, Furnace Cooled	22.4	119	122	0.3	7.3
As-received	15.7	117	122	2.6	7.2
950° C, Air Cooled	-23.1	140	186	30.0	8.9

*See footnote at beginning of Section 2 for composition of steel.

warmed to room temperature in cold water, and dried with alcohol and a warm air blast. The strain pattern was easily observed on the prepolished surface and was examined subsequently by optical microscopy. The plastic strain along the length of the specimen was determined by measuring the inscribed millimeter scale under a low-power microscope with a Hurlbut counter.

EXPERIMENTAL RESULTS

The stress-strain conditions. The preyield nonelastic phenomena and the initiation of Lüder's strain at -196 C in these specimens have been described elsewhere[12]. Typical load-extension curves are shown in Fig. 1. As anticipated[21], the yield stress was found proportional to the logarithm of the crosshead speed (v_c) (Fig. 2). The maximum drop in load during yielding in any specimen corresponded to a decrease of only 3500 psi, based on the original cross section. Contrary to the classical model of Lüder's-band propagation, the average strain over the length of the bands increased with the time of loading (Fig. 3). In addition, the local strain within a well-developed Lüder's band was not uniform (Fig. 4), it being a little greater near the point of initiation at the shoulder of the specimen. An average "true" stress was calculated from the measured load, and the corresponding average Lüder's strain at selected times. Although the load decreased somewhat as the Lüder's strain increased during the time-dependent yielding, the stress remained constant to within about 500 psi, which is an appreciably smaller

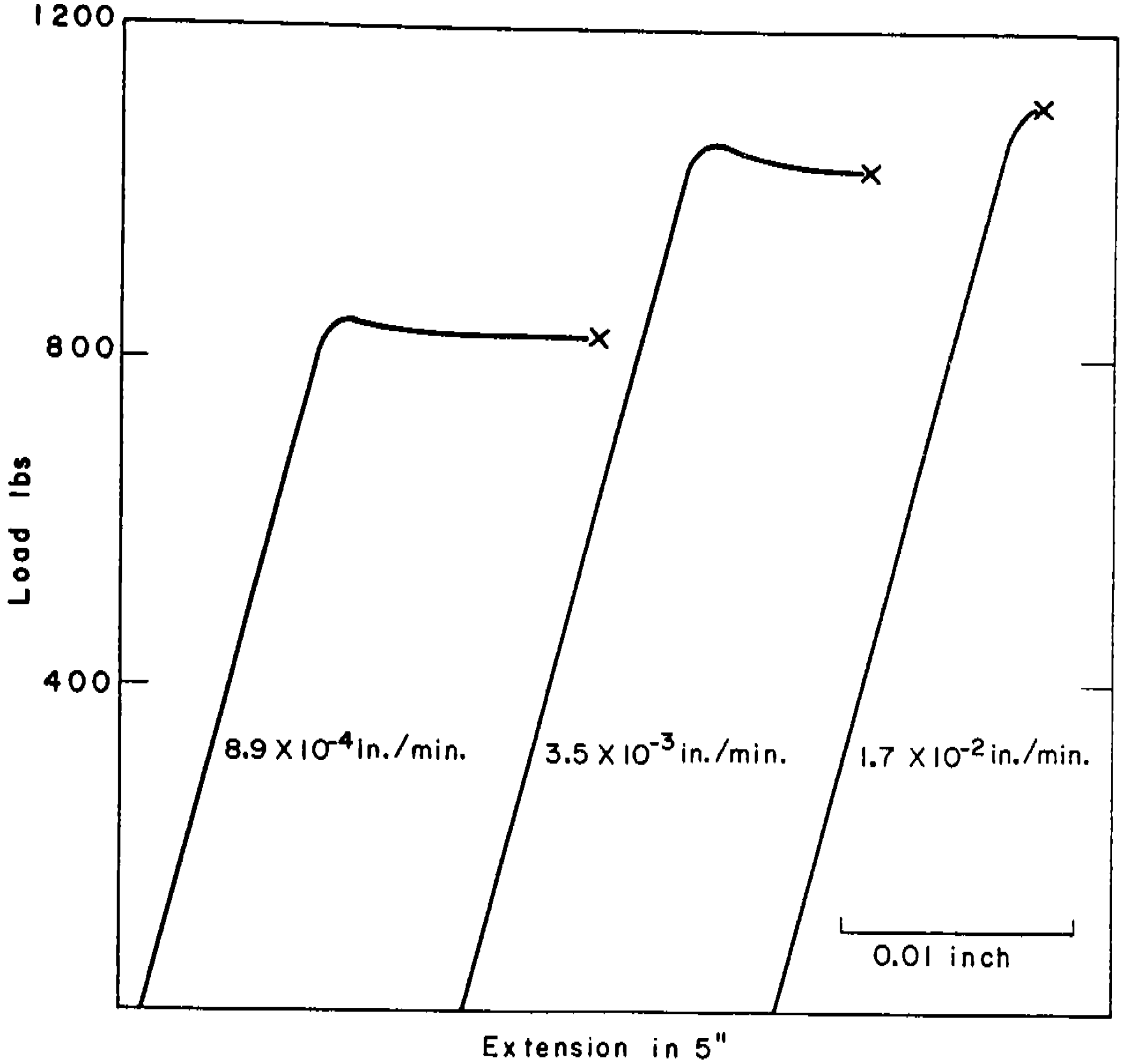


FIGURE 1. Typical load-extension curves at -196° C with different crosshead speeds. Steel as received.

X - cleavage fracture.

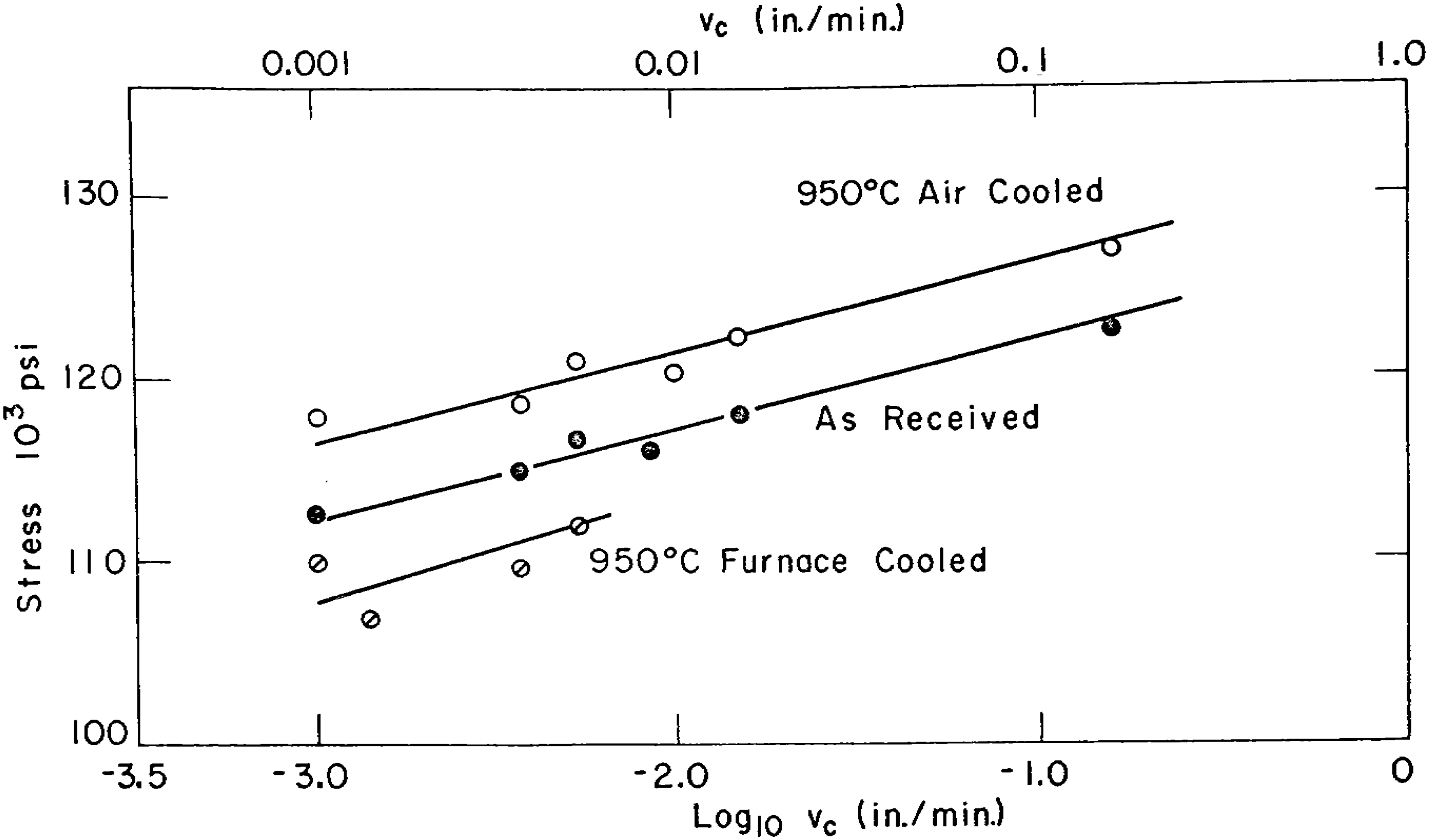


FIGURE 2. Dependence of upper yield stress on crosshead speed (v_c).
Tested at -196° C.

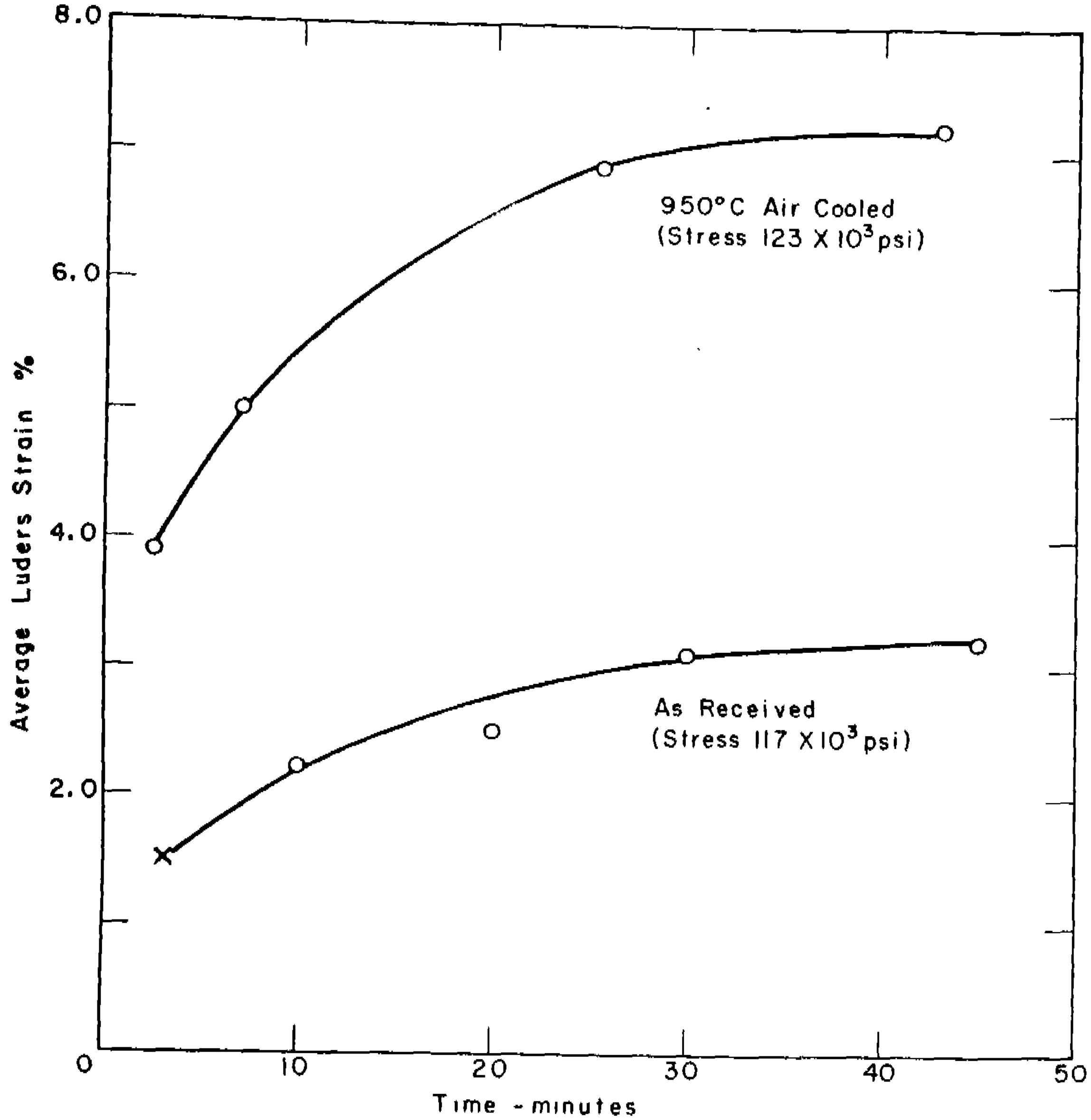


FIGURE 3. Average Luder's strain as a function of time of yielding. Steel tested at -196° C with a crosshead speed of 3.6×10^{-3} in/min.

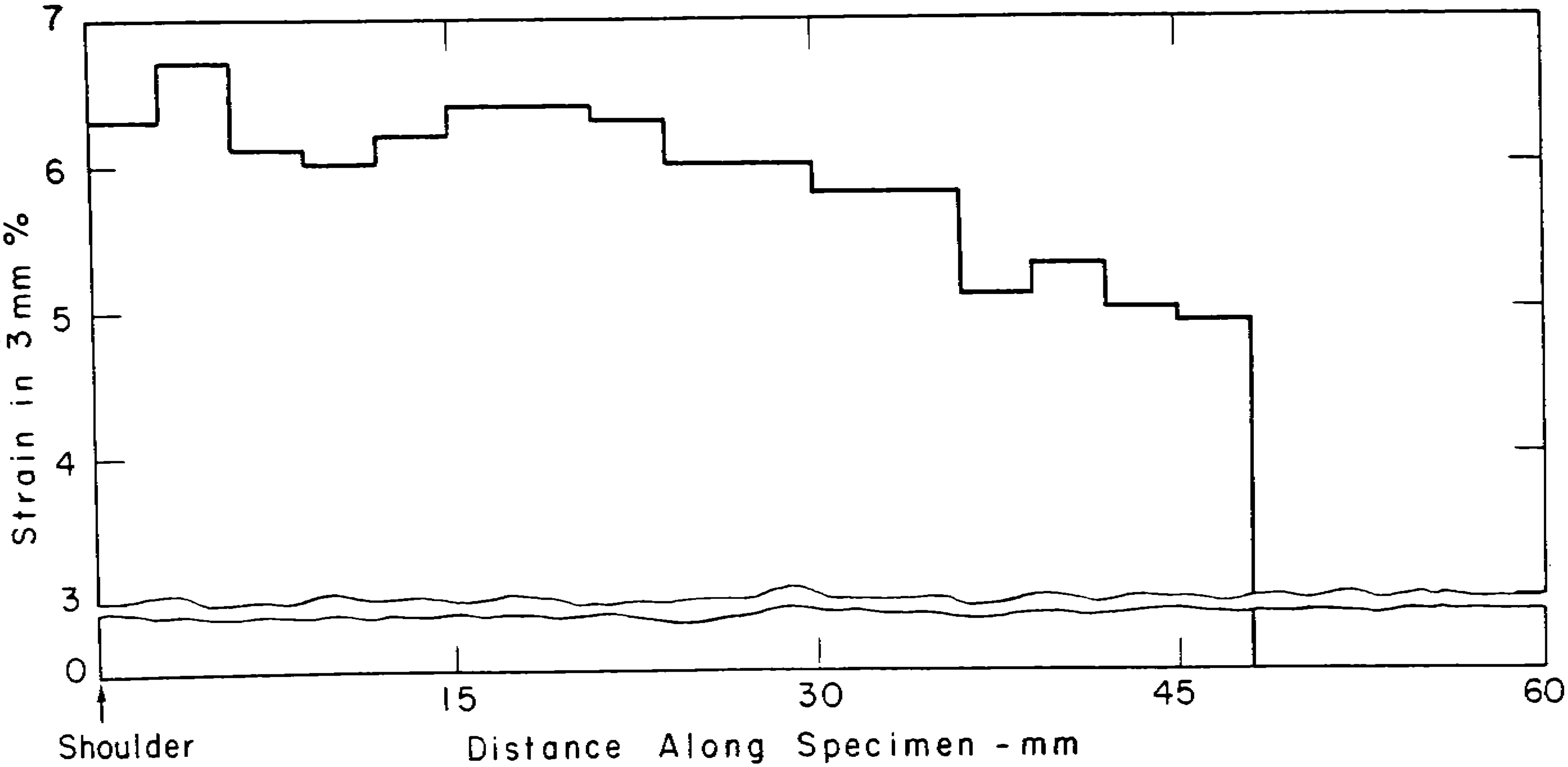


FIGURE 4. Example of distribution of strain in a Lüder's band when measured over 3-mm gauge lengths. Steel air cooled from 950° C. Tested at -196° C with a crosshead speed of 3.6×10^{-3} in/min.

variation than the experimental reproducibility of the yield stress (about 2000 psi).

Fisher and Rogers[14] have demonstrated the existence of creep at -78 C in mild-steel wires after the whole specimen is covered by Lüder's bands. The time-dependent Lüder's strain at -196 C observed in the present experiments appears to be a related effect. According to Fisher and Rogers[14], the Lüder's strain depends upon the applied stress and is not directly affected by the testing temperature. Unfortunately, the Lüder's strain is also influenced by the stiffness of the testing machine and the geometry of the specimen[15], so that it is not feasible to compare data obtained by different techniques. After normalizing at 950 C, there was a striking increase both in the creep rate and in the maximum strain (Fig. 3). At the same loading rate, the normalized specimen had a yield stress 6000 psi greater than the as-received specimen, but it seems improbable that this 5.1% increase in stress is sufficient to account for the 130% difference in Lüder's strain. Thus, it appears that at -196 C the Lüder's strain is sensitive to changes in heat treatment.

Lüder's bands started at both ends and moved slowly towards each other. Between the bands, the steel was stressed above the elastic limit and contained a small nonelastic microstrain (stage 1) which then increased slowly with time (stage 2)[12]. At very slow crosshead speeds, the maximum load could be applied for several hours before the Lüder's bands traversed the whole specimen. During this time, fine slip lines could be observed between the bands. However, no new Lüder's bands were initiated although it seems probable that this would occur with longer specimens or still slower loading rates.

Influence of loading rate on cleavage fracture. At constant stress, the time to fracture or, if no break occurred, to the termination of the test was measured from the onset of gross yielding as indicated by the load-extension curve. There was always a time delay before fracture which, as an example, is plotted in Fig. 5 as a function of the crosshead speed for the air cooled specimens. In this case, there was a well defined crosshead speed (8.5×10^{-3} in./min) above which the specimens fractured quickly and below which they were ductile. All the specimens furnace cooled from 950 C broke in a short time but showed a decrease in time-to-fracture with increase in crosshead speed. A similar situation is represented by the right-hand branch of the curve in Fig. 5. The slowest loading rate for brittle fracture of the as-received specimens lay between those for the other two series. At crosshead speeds greater than 3.6×10^{-3} in./min, all the as-received specimens broke after some delay. However, in seven tests at this border-line speed, three specimens broke quickly while the remaining four did not fracture in long-time tests.

The overall elongation (in 5 in.) was governed by the extent of Lüder's strain and the distance covered by the bands during the time of the test. The measured elongation is shown as a function of the crosshead speed in Fig. 6.

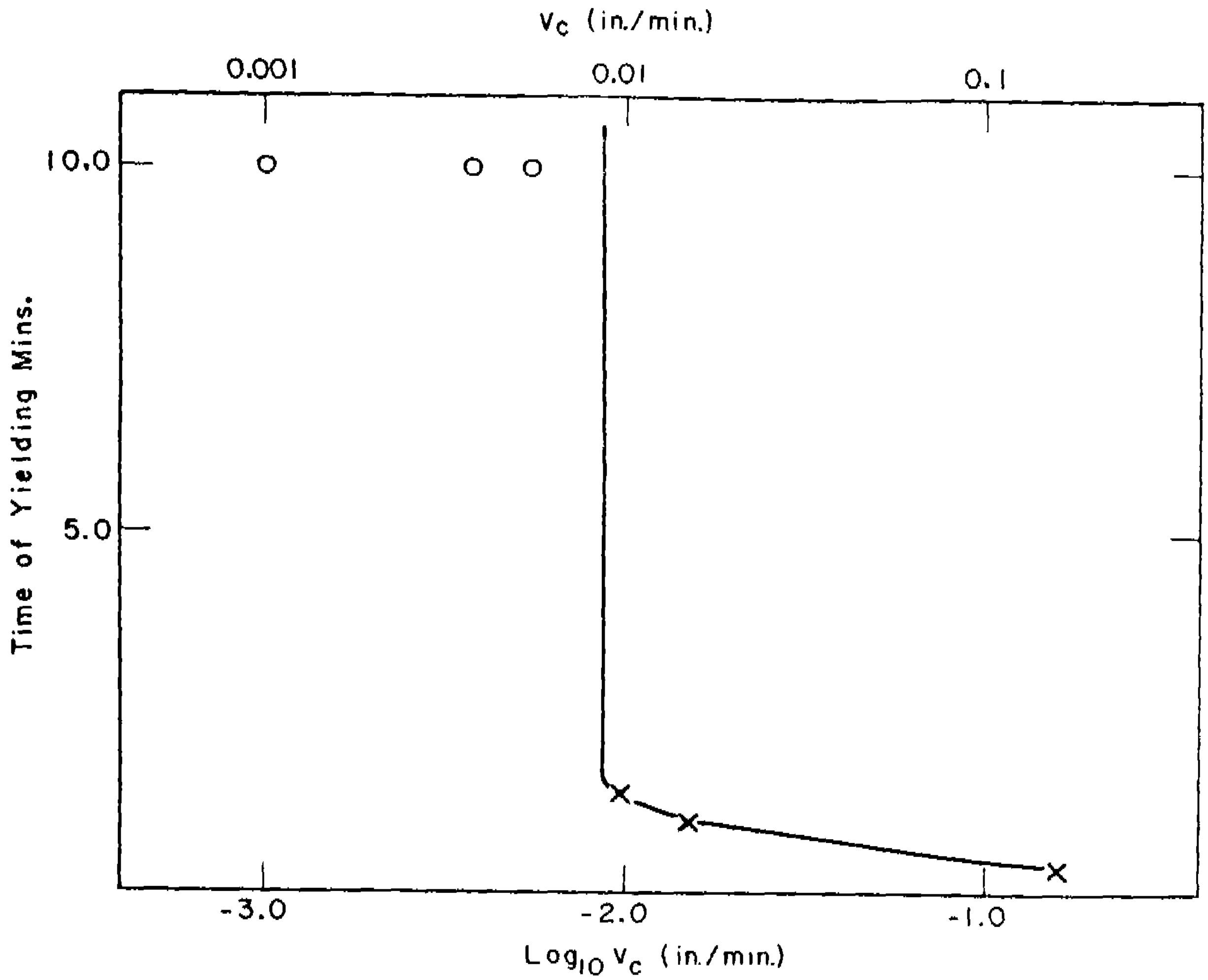


FIGURE 5. Time of yielding before fracture as a function of the crosshead speed (v_c). Steel air cooled from 950° C and tested at -196° C.

X - specimen broke
O = no fracture

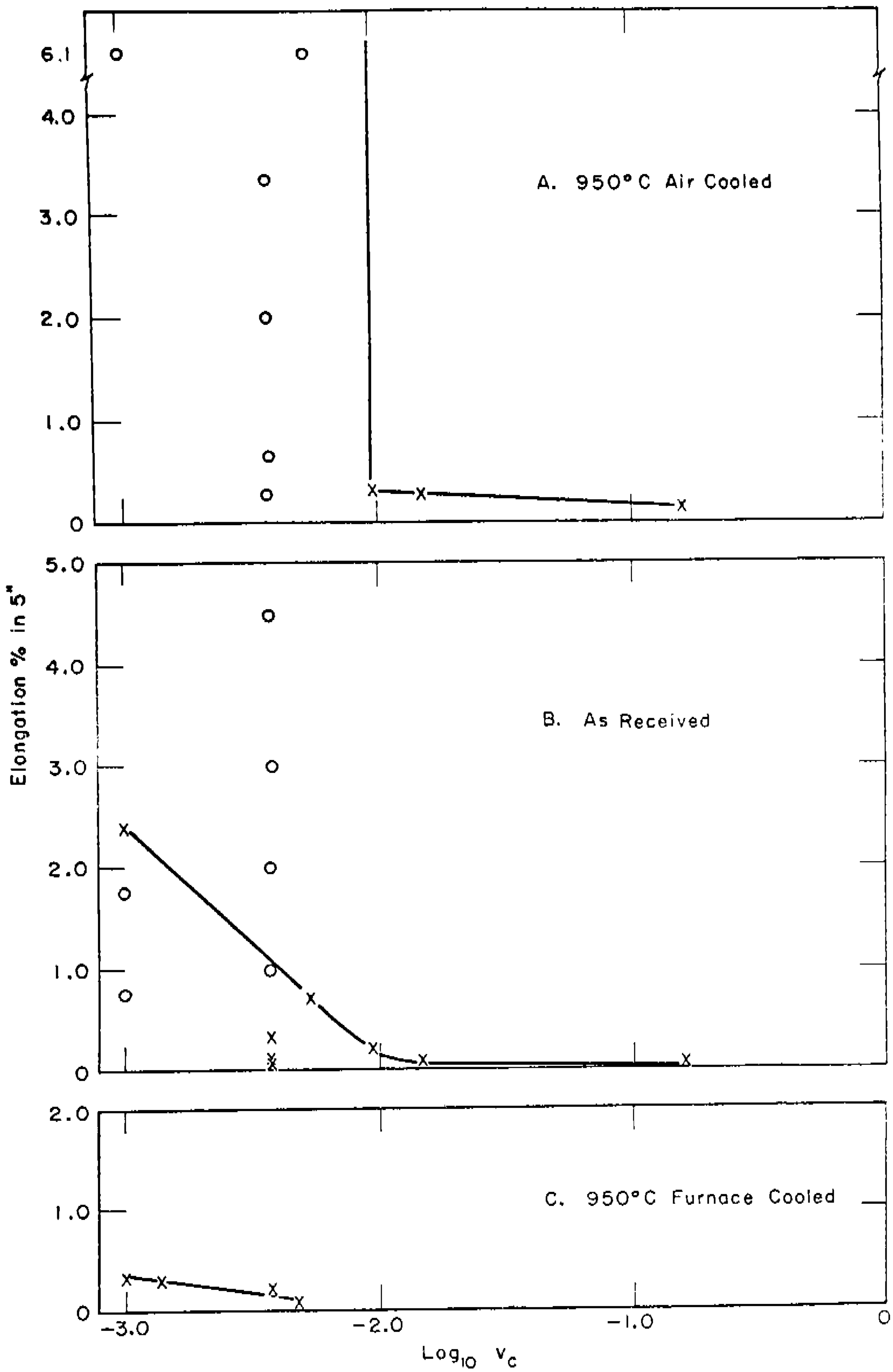


FIGURE 6. Elongation in 5-in. gauge length as a function of crosshead speed (v_c). Temperature -196° C.

x - specimen broke

0 - no fracture

When the overall elongation is plotted against the average "true" stress in the Lüder's bands at the end of each test, the curves shown in Fig. 7 are obtained. These curves suggest that in each series there is a stress above which the steel is brittle. In the normalized series the demarcation was well defined, but in the furnace cooled specimens a sufficiently low stress to permit ductile behavior was not attained. The lowest stress achieved experimentally for the as-received specimens was very close to the transition stress, and the scatter in the ductility data probably resulted from small differences in specimen dimensions or heterogeneity in the steel.

As the crosshead speed was increased in the brittle range, the yield and fracture stresses increased (Figs. 2 and 7) and the time to fracture decreased (Fig. 5). However, no fracture occurred without prior gross deformation. The average Lüder's strain in fractured specimens was between 2.1% and 5.2%, while the maximum strains in 3 mm were between 3.2% and 6%. Larger average strains were found in many of the ductile specimens. For example, average strains greater than 7% (with a maximum in 3 mm of 10.1%) were observed in normalized specimens tested at slow speeds such that no fracture occurred during the run (Fig. 3).

<u>Metallographic observations</u>. By visual examination of the polished surface, it was noted that each fracture was located within a Lüder's band.

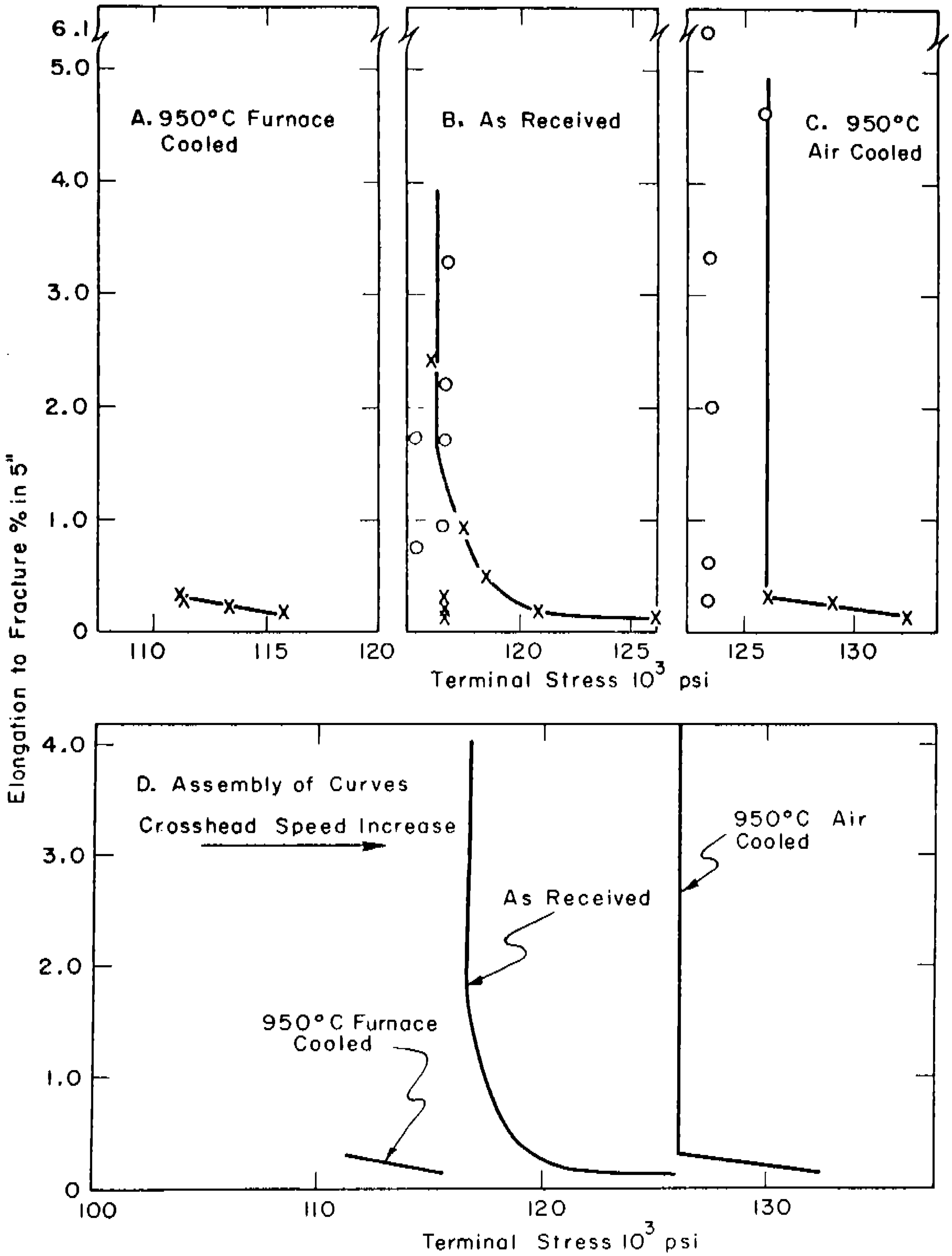


FIGURE 7. Elongation in 5-inch gauge length as a function of the true stress at the end of the test. Temperature -196° C.

X - specimen broke
O - no fracture

This was true even at the fastest crosshead speeds where the Lüder's bands spread only about 1 mm before fracture. Although the fracture was often fairly close to the leading edge of a band, it never coincided with the interface and in some instances was more than 1/4 in. behind the front.

Microcracks were observed in the Lüder's bands of all the specimens, even those which behaved in a ductile manner. Usually the cracks traversed only one or two ferrite grains (Fig. 8) although there were instances when several microcracks were connected by very severe local deformation to form an assembly that extended across five or more grains (Fig. 9). The longest strings of cracks were found in specimens subjected to the slowest loading rate and lowest stress. It appears that some single-grain microcracks spread slowly by inducing cleavage in neighboring grains. At low stresses this process is able to continue for some time before total fracture results.

A few twins were seen in the Lüder's bands, but these were easily distinguished from the microcracks both by their general appearance in rotated oblique illumination and by repolishing and re-examining the unetched surface when the cracks, unlike the twins, persisted. The twins were not associated with microcracks. In the area between the Lüder's bands no microcracks were noted in any of the specimens.

The genesis of the microcracks proved elusive. No microcracks smaller than a grain diameter were found, and all produced at their ends secondary deformation that tended to obliterate the original slip traces. In addition to the

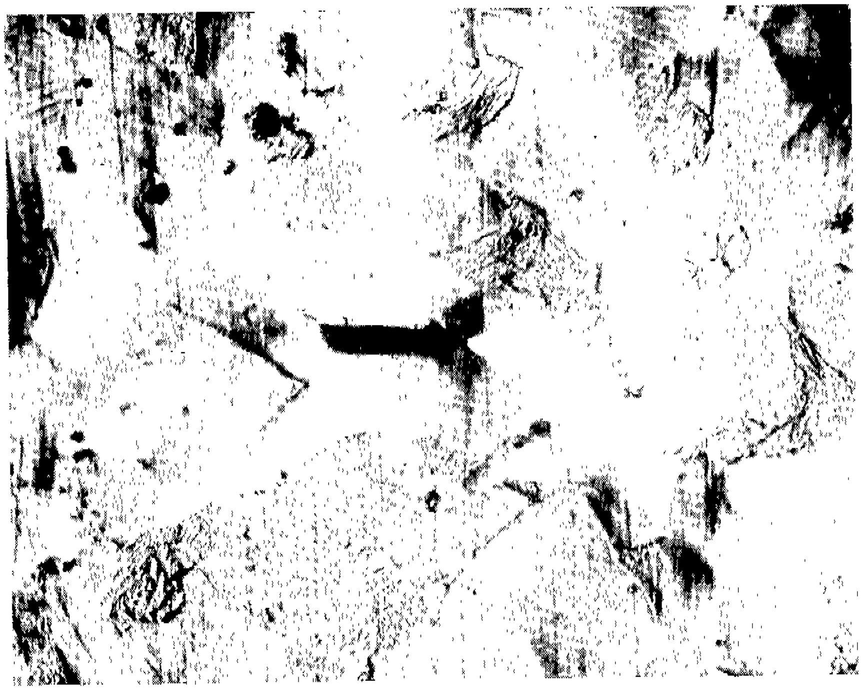


FIGURE 8. Examples of microcracks formed at -196° C in ductile specimens. X1000.

(a)

(b)

FIGURE 9. Assembly of microcracks in specimen without a complete fracture. (a) and (b) the same area. (a) X150. (b) X1000.

microcracks, numerous markings were observed that indicated slip at variance with the predominant slip orientation and with exceptional step heights (Fig. 10). However, no direct evidence connecting these with the subsequent formation of microcracks was obtained.

DISCUSSION

Comparison of the Stroh and Low theories. Stroh[1, 2] has postulated that a cleavage crack can form when n dislocations, piled up under the action of a shear stress, σ_s, satisfy the condition

$$nb\sigma_s = 12\gamma \tag{1}$$

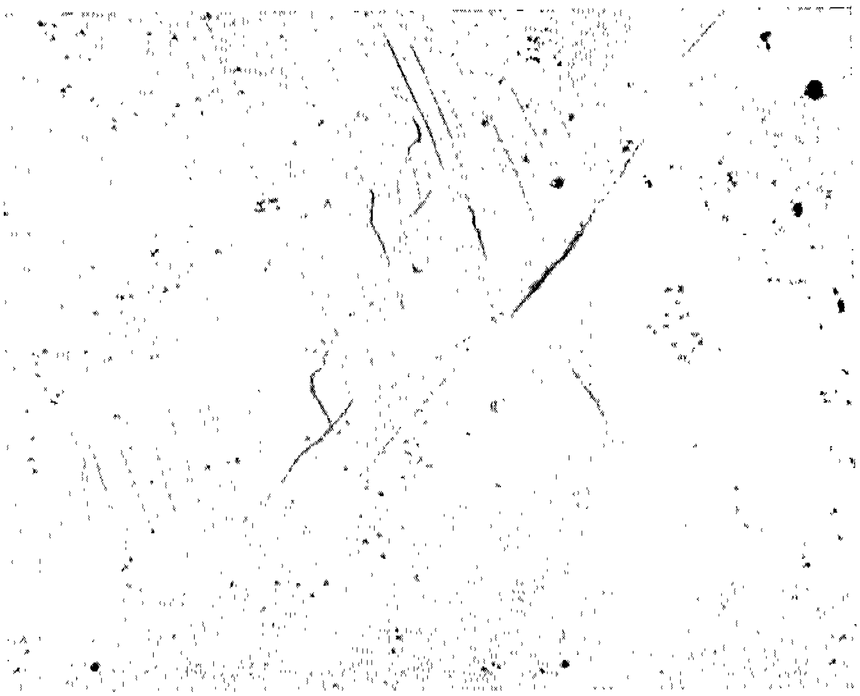
where b is the Burger's vector and γ the surface energy of the cleaved faces of a single crystal. Assuming the blockage is provided by a grain boundary, the piled-up dislocations are pictured to occupy a length equal to half the grain diameter d. The fracture stress σ_F is then

$$\sigma_F = 4\left[\frac{6\gamma G}{\pi(1-\nu)}\right]^{1/2} d^{-1/2} \tag{2}$$

where G is the shear modulus and ν is Poisson's ratio. At a given temperature,

$$\sigma_F = kd^{-1/2} \tag{3}$$

which has been verified experimentally for polycrystalline zinc by Greenwood and Quarrell[5]. To fit the experimental data for iron and steel at -196 C, Petch[3, 4] found it necessary to introduce a second constant σ_o, the yield



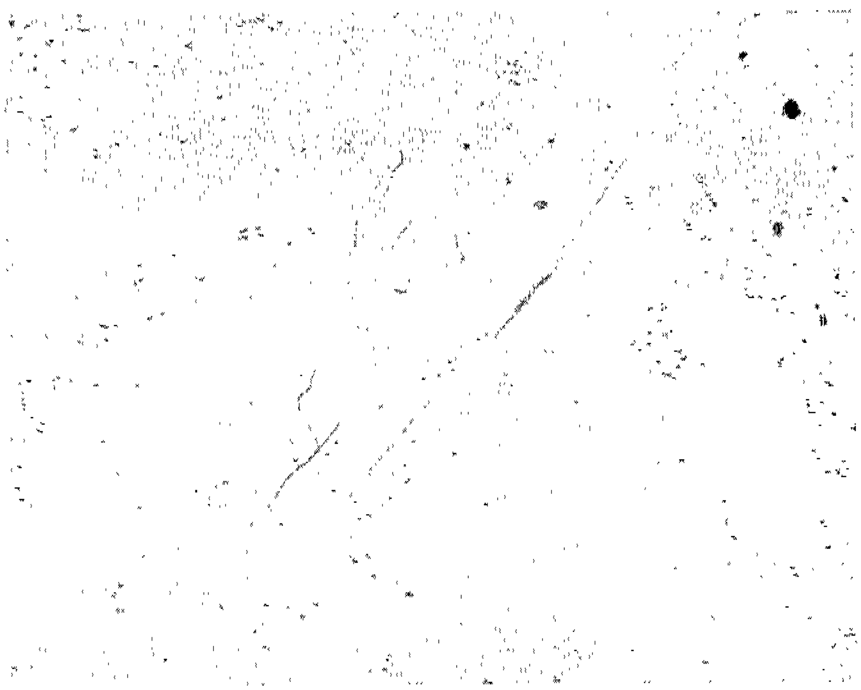


FIGURE 10. Unusual slip markings observed after deformation at $-196°$ C. (a) and (b) are the same area with different positions of the oblique illumination. X1000.

stress of a single crystal at the testing temperature, so that

$$\sigma_F = \sigma_o + kd^{-1/2} \tag{4}$$

Low[10] also observed a linear relationship between σ_F and $d^{-1/2}$ for low-carbon iron at -196 C but explained the result in terms of the Griffith-Orowan concept:

$$\sigma_F = \left[\frac{4Ep}{\cdot\cdot}\right]^{1/2} d^{-1/2} \tag{5}$$

where E is Young's modulus and p is an energy term which includes the surface energy and the plastic work done during fracture.

According to Stroh[2], the proportionality constant k in Eq. (3) (which does not include a plastic-work term) agrees with experimental values deduced from the Petch[3, 4] and the Greenwood-Quarrell data[5]. On the other hand, Low[10] maintains that the plastic-work term far outweighs the surface-energy term since the experimental value found for p in Eq. (5) is at least two orders of magnitude greater than the known surface energies of metals.

An important stimulus to the development of the dislocation theory has been the belief that subcritical Griffith cracks do not exist in a real metal prior to brittle fracture. The present work has shown clearly that this belief is unfounded in the case of mild steel since microcracks are formed before fracture.

The dislocation pile-up theory does not account for either the location of the fracture or the effects of loading rate observed in the present experiments. In this view[2], fracture and yielding are taken to be competitive processes, cleavage resulting from the normal stresses and yielding from the shear stresses around a queue of dislocations. The shear stress required to activate a dislocation source in a neighboring crystal increases as the temperature is decreased in such a way that at some temperature the necessary normal stress for cleavage is achieved before the shear stress to induce further slip is reached. The present experiments were carried out well below Stroh's estimated temperature[2] of transition from slip to cleavage behavior, and also below the measured tensile transition temperature[16]. Consequently, the strain preceding fracture should have been only that resulting from the pile-up of dislocations, in magnitude about the same as that of the first-stage microstrain. In other words, brittle fracture should have occurred either before gross yielding in the vicinity of the stress concentration near the specimen ends or, subsequently, at the front of the Lüder's bands. In none of the experiments did fracture occur in either of these situations.

The time required to build up a dislocation queue at a constant applied stress should be that necessary to establish the microstrain (stage 1). These times for mild steel, at different stresses and temperatures, have been measured by Clark and Wood[17--19]. At the static yield stress they are very small compared with the delay time for yielding. Thus, since there is no term in Eq. (2) suggesting a built-in time dependence for cleavage, the Stroh theory predicts that the fracture stress should be less sensitive than the yield stress to an increase in the loading rate and that the probability of fracture without prior gross yielding should be greater when the crosshead speed is increased. In the present experiments fracture was always preceded by gross yielding and the fracture stress increased with the yield stress.

The Low concept of brittle fracture in iron at -196 C rests on two propositions:

1. Fracture is preceded by yielding and by the formation of microcracks of a size related to the grain diameter.
2. The fracture stress is that stress which causes the microcracks to propagate to complete failure, the critical conditions being defined by Eq. (5).

The relation between yielding and fracture was demonstrated experimentally by showing that in coarse-grained iron the fracture stress in tension coincides with the yield stress in compression when both tests are conducted at -196 C and at the same strain rate. A separate slow-bend test was used to demonstrate the presence of subcritical microcracks. These microcracks were caused to propagate to complete failure by increasing the applied stress. Thus, the conditions were comparable with those in the fourth stage of the nonelastic strain in a tension test; that is, after the Lüder's bands have covered the specimen.

The present experimental data, which show this behavior even in the situation of minimum ductility in tension at -196 C, provide substantial support for Low's viewpoint over a range of testing speeds. At all stresses which produced gross yielding at -196 C, microcracks were found in the Lüder's bands. In the air cooled and as-received specimens, these did not propagate to failure within the time of testing (about 40 min) unless the stress was greater than a certain minimum (Fig. 7D). It is reasonable to assume that in the more brittle furnace cooled specimens the yield stress was always greater than the minimum stress required to cause the microcracks to spread rapidly.

<u>General discussion</u>. To form a microcrack it is necessary to cause appreciable plastic deformation and to reach a sufficiently high stress. The minimum stress necessary at -196 C could not be ascertained since microcracks were produced at the lowest yield stress attainable. Nor was it possible to obtain any direct information about the mechanism by which the microcracks are generated. Single-crystal techniques appear to be more appropriate for studying this problem.

At a constant applied stress, there is a time delay before cleavage fracture (Fig. 5). This has three components: the delay time for the initiation of Lüder's bands, the time for formation of microcracks, and the interval between microcrack formation and propagation to complete fracture. The first component is not significant in the present data since the times were measured from the start of gross yielding. In the slow-speed tests on air cooled specimens, microcracks were found within 30 seconds after the initiation of Lüder's strain, although the bands continued to spread for an additional 40 minutes without fracture. Evidently, the propagation of the cracks is the major time-deciding factor.

The delay in propagation cannot be understood in terms of Eq. (5) which does not contain any time-dependent factors. The average plastic strain (although time-dependent) is not a criterion for fracture since the slowly loaded specimens that did not break contained a much larger Lüder's strain than the specimens that fractured at higher stresses. As explained earlier, the average stress on a Lüder's band did not fluctuate more than about 0.4%. While it is possible that this fluctuation could account for the delay in fracture, it seems more probable that the important changes occur on a much smaller scale.

The delayed fracture is thought to be a consequence of the time dependence of the Lüder's strain at low temperatures (Fig. 3). As measured on the millimeter grid, the strain along a band is not uniform (Fig. 4), and frequently fractures are not located near the maximum strain even on this scale. Microscopically the deformation is very heterogeneous. The stress-strain pattern in the vicinity of any one grain is not accurately reflected by the average measurements, and it changes continuously as creep progresses. Under these conditions the development of microcracks is complex. The material surrounding a crack is not isotropic, as assumed by Eq. (5). Further, the relative grain orientations must be a factor in determining the probability that one of the original microcracks will induce cleavage in a neighboring grain. When cracks have formed in several connected grains, d in Eq. (5) should be replaced by the length of the crack assembly, and the same applied stress can now drive the fracture through more unfavorable environments. There is some critical size such that fracture can progress through any surroundings it encounters, and then catastrophic failure ensues.

A sequence of cracks extending through a number of grains, N, (Fig. 9) was observed in several of the specimens that had yielded extensively at stresses below the minimum stress for fracture. Similar, but less extensive, arrays of cracks were observed in some of the fractured specimens in locations remote from the main fracture. It is evident that, at stresses close to and below the minimum fracture stress, the critical size for final failure is appreciably greater than that for the first-formed microcrack. If Eq. (5) is used to compute the minimum fracture stress, d must be replaced by N*d, N* being the number of single-grain microcracks in the assembly of critical size. Then,

$$\sigma_F = \text{Const.}\left(\frac{p}{N^*d}\right)^{1/2} \qquad (6)$$

The present experimental conditions approximate the situation where the applied stress σ is held constant while N increases with time.

Fracture occurs if the total crack length (Nd) reaches a critical value in the time available. It may be that the growth rate of a sub-critical crack is affected by the grain size and heat treatment, but no information on this point is available. However, it is clear that the critical crack length is determined by the value of p. Grain size is an important factor in fracture behavior and, thus p must be a function of d. This is readily understood if the contributions to the plastic work term are considered. The two major sources of energy absorption are the deformation which occurs at cleavage steps within the grains[9, 20] (p_{step}) and the shear deformation required to join adjacent cleavages (p_{tear}). The latter affects a large volume of metal (Figs. 8 and 9) and certainly appears to be the major contribution. Since the number of discontinuities that occur during cleavage, and therefore the number of sheared volumes, must increase with a decrease in grain size, it seems likely that the grain size effect operates by viture of its influence on p_{tear}, and thus on p.

It has been suggested that σ_F is proportional to $d^{-1/2}$ [3, 4, 9, 10], but the minimum fracture stress as revealed by the present experiments conforms only very approximately to this relation (Table 2). The range of d investigated was small, but appreciable effects were produced by

TABLE II

Variation of $\sigma_{F_{min}} d^{1/2}$ with Heat Treatment

Heat treatment	$\sigma_{F_{min}} d^{1/2}$ (lb/in.$^{5/2}$)
950 C, furnace cooled	<3.97
As-received	4.18
950 C, air cooled	3.22

changes in heat treatment. For example, at 125×10^3 psi the furnace cooled and as-received specimens fractured in a short time, while the air cooled specimen was ductile (Fig. 7D). This behavior reinforces a previous conclusion[13] from work on these steels that changing the cooling rate produces greater effects than can be accounted for on the basis of grain size alone.

Low[10] and Petch[4] employed short tensile bars of circular cross section and probably loading rates‡ approaching the most rapid used in the present experiments. Under their conditions, the creep and delayed fracture phenomena are not significant because the specimen is at the constant lower yield for only a short time. As stated by Low[10], in coarse-grained steel microcracks form and propagate to failure at the yield stress, but in fine-grained steel the microcracks do not propagate until the continuously increasing applied stress reaches some higher value. In the former case, σ_F closely corresponds with σ_{YS} and both are the same linear function of $d^{-1/2}$. The second situation, when the specimens show some ductility, is complex. Then the first-formed microcrack (about one grain diameter) is smaller than the critical size (N^*d) and, consequently, it is expected that the size of the crack assembly (Nd) grows as the stress (σ) is increased. It is expected that p also will increase with σ since further plastic deformation

‡Their loading rates are not given.

occurs. Low found experimentally that the additional stress required to cause crack propagation to failure was proportional to $d^{-1/2}$ although the data indicated appreciable scatter, and subsequent investigations have shown that grain size is not the only heat-treatment-sensitive variable in this situation[13, 22].

CONCLUSIONS

The following refers only to mild steel, long-strip specimens, tested in tension at -196 C with testing speeds between 8.9×10^{-4} and 1.6×10^{-1} in./min.

1. Fracture is always preceded by gross yielding; thus, when the testing speed is increased, the fracture stress increases with the yield stress. Microcracks and complete fractures are located only in the Lüder's bands, usually some distance behind the moving front.

2. Although microcracks are formed at all loading rates within the stated range, there is a minimum stress required for complete fracture. This minimum stress is materially affected by heat treatment.

3. The magnitude of the Lüder's strain is time dependent and sensitive to changes in heat treatment.

4. Fracture occurs in stages: first microcracks, about one grain diameter, are formed. By inducing cleavage in neighboring grains, these microcracks spread subsequently across a number of grains, so that the

individual cleavages are connected by regions of very severe deformation. Finally, the assembly of cracks reaches a critical size and propagates rapidly to failure.

5. At a constant stress during yielding, there is a time delay to fracture. This appears to be a result of the creep which occurs in the Lüder's strain.

In general terms, these observations are compatible with Low's application of the Griffith-Orowan concepts rather than with the dislocation pile-up theory. Thus the latter does not appear applicable to the brittle fracture of mild steel.

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