SSC-162

Exhaustion of Ductility and Brittle Fracture of E-Steel Caused by Prestrain and Aging

by

C. MYLONAS

SHIP STRUCTURE COMMITTEE

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ADDRESS CORRESPONDENCE TO:

SECRETARY
SHIP STRUCTURE COMMITTEE
U. S. COAST GUARD HEADQUARTERS
WASHINGTON 25, D. C.

July 1964

Dear Sir:

In order to study the effect of gross strain upon the mechanical and metallurgical properties of steel and to relate these variables to steel embrittlement, the Ship Structure Committee is sponsoring a project at Brown University entitled "Macrofracture Fundamentals." Herewith is a copy of the Second Progress Report, SSC-162, Exhaustion of Ductility and Brittle Fracture of Project E-Steel Caused by Prestrain and Aging by C. Mylonas.

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

Comments on this report would be welcomed and should be addressed to the Secretary, Ship Structure Committee.

Yours sincerely,

T. J. Fabik

T. J. FABIK
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

SSC-162

Second Progress Report
of
Project SR-158
"Macrofracture Fundamentals"

to the

Ship Structure Committee

EXHAUSTION OF DUCTILITY AND BRITTLE FRACTURE OF E-STEEL CAUSED BY PRESTRAIN AND AGING

by

C. Mylonas
Brown University

under

Department of the Navy
Bureau of Ships Contract NObs-88294

Washington, D. C.
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ABSTRACT

The investigation of static brittle fracture initiation in engineering structures requires first the establishment of a criterion of brittle behavior of the structure as a whole. Such a criterion is obtained by a comparison of the fracture load with the flow limit of an idealized perfectly plastic material. The difference between static fractures at high and low load was related to the magnitude of the plastic strains at regions of strain concentration and to the ductility of the steel. The contrast between static laboratory tests of notched plates of sound steel which did not fracture before the flow limit was reached, and service failures which have occurred at low nominal stress levels, shows that the ductility of sound steel is sufficient to avoid low average stress fracture, but may be reduced during fabrication or in service. This was demonstrated experimentally with extensive tests of pre-strained notched plates, bent bars, and axially compressed bars. It was found that the ductility depends on the whole history of strain and temperature and is suddenly and drastically exhausted by cold straining of a closely determined amount, and far more easily by straining at about 500 F. This led to the first systematic static brittle fracture initiation of unwelded steel plates at low average net stress, as low as 10% of yield. These results provide an explanation of the initiation of service failures, which are usually traced to cold worked regions or to defects close to welds, where complex hot straining occurs. Further tests have shown that the ductility of cold strained steel is restored by a heat treatment at about 1100 F or higher. The required duration of heat treatment is shorter for hot than for cold-strained bars and appears to increase with the amount of prestrain, and to decrease when the temperature is raised. A better understanding of the mechanism of fracture initiation makes it now possible to express qualitative macroscopic criteria of fracture based on the strain hardening law and the ductility of embrittled steel and on the strain and stress distribution at a sharp notch in such material.

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1. GENERAL CONCEPTS

Several reports and review articles describe the conditions under which service failures have occurred, and the practical measures which led to a reduction of their occurrence. The present report is concerned with the mechanism of fracture initiation in structures subjected to static loading, and is based in part on earlier research (1)-(12),* carried out at Brown University under the sponsorship of the SSC.

1a. Distinguishing features of brittle fracture. The first difficulty encountered in the study of brittle fracture of steel structures is the lack of a clear definition. Simplified definitions of brittleness of materials are not applicable. A cleavage appearance is not a requirement of brittleness in the failure of steel structures, which always exhibit a good measure of shear fracture mixed with cleavage. Nor is a complete absence of ductility a good criterion, because even the most brittle service failures show signs of plastic deformation at the point of initiation. Likewise the local stress at the point of initiation is always high since fracture starts at defects or points of stress concentration. The problem of brittle failure becomes clearer only when the interdependence of local and overall behavior of the structure as a whole is considered. The laws of plastic deformation and limit analysis of structures have proved invaluable in this respect.

The characteristic feature of brittle failure of structures was indicated by the difference between service failures and laboratory tests. In at least a few clear-cut instances service failures occurred under static loading at overall stress levels well below yield (13), and lower than nominal stress levels successfully sustained in similar structures. It is believed that this is true in the majority of service failures. On the contrary, it had not generally been found possible to reproduce static low-stress brittle initiation of fracture in the numbeous early laboratory tests. Contrary to the elementary Griffith-type theories of fracture (14)-(16) symmetrically notched plates of mild steel having even the deepest and sharpest cracks and temperatures below brittle transition, were not found to fracture in central static loading before the average stress over the net section reached yield level and appreciable plastic deformation occurred (1)-(12). This difficulty or barrier to the static initiation of fracture in sound steel (1)-(4), (6), (11), (12), (17), could only be overcome by a strong impact at a notch, frequently in combination with local severe cooling or by fracture initiation at a brittle bead weld (19). Once started, however, the fracture would propagate at high speed in regions of low stress and higher temperature, just as in service failures.

It appears that brittle fracture passes through the two distinct phases of initiation and propagation. The inability of energy theories of the Griffith type to describe the initiation of fracture is not surprising, because they only express necessary and *not sufficient conditions.* When no other conditions need be satisfied, as in glass, necessary conditions are also sufficient. But the barrier to fracture initiation indicates that some other criterion apparently more stringent than energy balance must also be fulfilled, *probably a maximum stress* or strain criterion of fracture. Energy theories may be applicable to the stages of propagation or arrest of a crack, provided the dynamic effects are properly considered.

1b. Definition of Static Brittle Fracture Initiation. The fracture of a structure or structural member containing defects or notches, at a static load causing general yielding of the net section (i.e. at the flow limit) is not surprising or irregular. But fractures at static loads below the flow limit are certainly irregular, and in engineering will be called brittle. This definition may also be derived from a consideration of the magnitude of the local strains prior to fracture in relation with the overall loads and deformations. As discussed by Wells (20), plastic strains at the notch roots are relatively small and contained within elastic regions, as long as the average stress level is lower than the flow limit. They can become very large only when the flow limit is

*Numbers in brackets refer to the list of references at the end of the text.

reached and general yielding occurs. The discussion has tacitly referred to perfectly plastic materials for which these concepts are clear. The picture changes little with materials like mild steel. Even though strain hardening sets in at some stage, it is still found that the overall deformations increase distinctly more rapidly at a specific value of the load which usually is close to the flow limit of an idealized perfectly plastic material. The local strains will again increase slowly at first and rapidly after this limit, with an intermediate gradual transition. The basic idea is that in sound mild steel sufficient strain hardening and fracture can occur only at the high strains associated with the flow limit. This has been extensively discussed by Drucker (1) at the 1953 Conference of Brittle Fracture held at MIT. He concluded that low average stress fractures occur at small local strains, and conversely that a ductility smaller than needed to permit yielding at the notches up to the flow limit, will result in low average stress initiation of fracture. Accordingly the smallness of the average stress at fracture is an adequate and sufficient criterion of brittleness of fracture initiation in a structure. If the average net fracture stress is of yield intensity or higher, the fracture is ductile; if it is decidedly smaller the fracture is brittle. The average net stress however is not the direct cause of failure but only a convenient indication of the magnitude of the strains. No confusion should be made with the true peak stress at a crack or notch, which is of yield or raised yield intensity long before general yielding occurs. Fracture starts and advances in a local field of high stress.

The average net stress criterion of brittle fracture initiation has been the basis of the extensive research sponsored by the Ship Structure Committee at Brown University since 1954 (2)-(13).

1c. Causes of Static Brittle Fracture Initiation in Structures. The proposed criterion of brittle fracture initiation and the concepts on which it is based provide a clear understanding of the causes of brittle failure of structures. Although the material criterion of fracture is not known, it is clear that brittle fracture does not occur when the material has sufficient ductility under the conditions of stress and constraint existing at a notch, so as to be able to yield up to at least the flow limit. Therefore the failures obtained in the laboratory at average stress levels of yield intensity, even with the poorest structural steels under the worst conditions of stress concentration and below the transition temperature show that mild steel in its initial state has all the necessary ductility to avoid brittle fracture. Since low stress failures did occur in service it must be concluded that in the region of initiation the original ductility had somehow been reduced or exhausted during fabrication, service or repair.

It must also be concluded that the flow limit fractures generally obtained in the laboratory tests did not reproduce the phenomenon of static brittle fracture initiation. Such reproduction of low static stress failures is an indispensable step in the study of brittle fracture, and a fundamental check of the correctness of the proposed concepts. It was the first aim of this research. According to these concepts low stress initiation under static loading should become possible when the initial ductility of mild steel is sufficiently reduced, as e.g. by suitable cold working or heating. This conclusion was completely substantiated by a series of tests described in later chapters. Typically brittle fractures or arrested cracks were produced in unwelded precompressed notched steel plates, at stress levels as low as 10% of yield (Fig. 1). A drastic exhaustion of the original ductility was produced with sufficient axial compression, or bending of cold or hot bars, which subsequently fractured at extensions of the order of 2%. In general it was found that the remaining ductility depends strongly on the whole history of strain and temperature, and may be highly anisotropic.

These results, which are fully discussed later, substantiate the concepts of plasticity on which this investigation is based. They also clarify some unsuspectedly strong or unknown effects and properties of steel. They also indicate methods for studying the embrittling influences, and for selecting steels according to the properties which are important in fracture, i.e. the properties of the damaged steel.

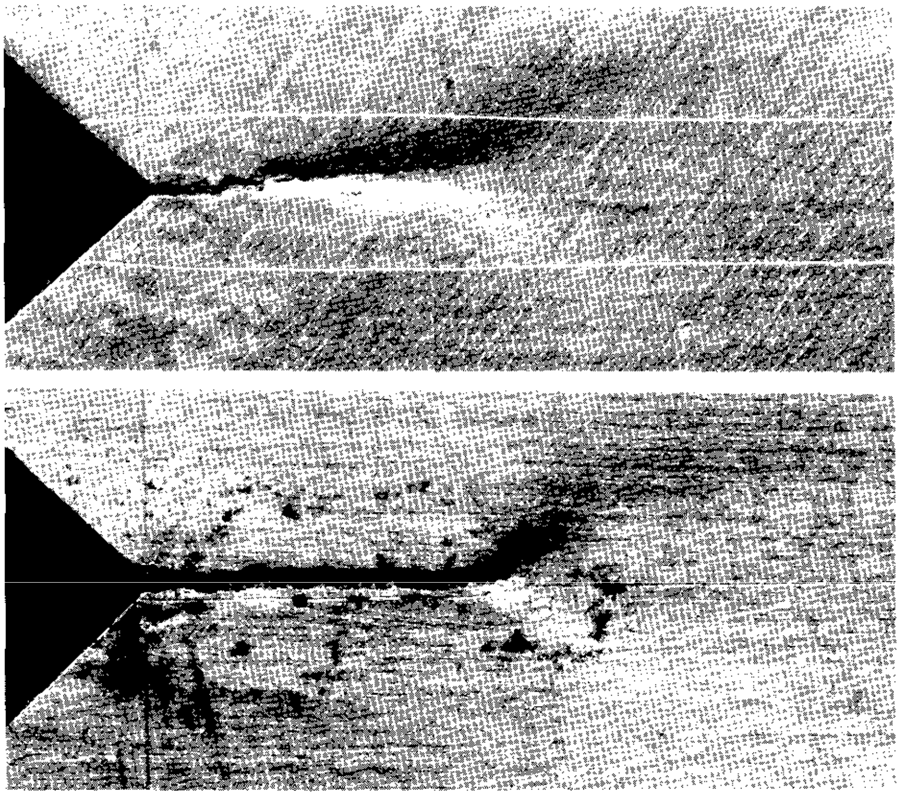


FIG. 1. NOTCH REGION DETAIL OF 10" SQUARE, 3/4" THICK NOTCHED PLATES PRECOMPRESSED AND TESTED IN TENSION VERTICALLY LOW LOAD BRITTLE INITIATION AND DUCTILE ARREST OF CRACKS.

Several independent investigators (21)-(41), starting from various points of view have studied many aspects of the problem of work hardening or reduction of ductility by straining at various temperatures. Notable among them is the work of Körber, Eichinger and Möller (21) at the Kaiser-Wilhelm Institute in 1941-43, which appears to have escaped the attention of all subsequent investigators, in the U. S. and abroad.* Our understanding of the mechanics of brittle fracture has been greatly delayed by this oversight.

1d. The Criterion of Fracture. Once low stress brittle fracture initiation has been consistently produced, and the importance of exhaustion of ductility demonstrated, the obvious aim is to find methods of assessing the danger of fracture. Fracture of the material may obey a strain criterion (e.g. a maximum strain under certain conditions of constraint), and then the danger of fracture should be assessed by a comparison of available material ductility and required ductility, under the local constraint existing at a notch, and for the strain hardening law of the specific material. The criterion may also be a maximum stress, which may be reached after sufficient work hardening, such as occurs with large strains, and under sufficient constraint against low stress yielding. But extensive ductility means slow work-hardening and a weak lateral constraint, hence a small stress. Thus even if fracture obeys a maximum stress criterion, its fulfillment depends on the magnitude and type of the strains, as well as on the complete strain hardening law. Provided the interdependence between ductility, strain hardening, constraint, and stress is understood and the exact stress-strain relations and strain distribution are taken under consideration, the conditions of fracture should be equally well expressed in terms of stress or strain. At the moment there is no way of knowing which expression will be simpler or more realistic.

*This work has been brought to our attention by Professor N. H. Polakowsky of the Department of Metallurgy, Illinois Institute of Technology.

The answer to the problem of brittle initiation requires a knowledge not only of the exact strain-hardening law and of the remaining ductility after a damaging strain history, but also of the true strain distribution at a crack or notch. Studies of the properties after various types of straining are reported in the chapters which follow. The problem of the true strain distribution at a notch in a strain hardening material such as prestrained steel is quite difficult. Stress distributions fulfilling differential equation of equilibrium, compatibility and boundary conditions have been given for a notch in a perfectly plastic material in plane strain (42)-(43), but the proof of uniqueness is lacking. An exact solution has also been given for a notch in plane strain subjected to shear (44)-(45). No solution exists for a plate of finite thickness which has a three dimensional distribution of stress, but it has been shown that conditions of plane strain are approached only when the thickness is many times larger than the width of the net section (46). Plane strain conditions in elasticity are reached at a much smaller thickness, which need only be large in comparison with the radius of the notch root (46)-(47).

Attempts have often been made to obtain approximate solutions of the stress distribution around notches in plane strain with the help of gross simplifications and arbitrary assumed distributions. For example in a recent attempt (48), use is made of the elastic solution around a hyperbolic notch in plane strain. The plastic zones are assumed to be regions where the elastic strains violate the Mises yield condition. The plastic stresses are taken to vary linearly from the elastic plastic interface to the notch root, and the relative elastic stress distribution in the elastic region is assumed unchanged. An adjustment is made to achieve overall equilibrium between external load and the altered stress distribution, but otherwise neither differential equations of equilibrium, nor compatibility, nor boundary conditions are satisfied anywhere in the plastic zone or at the elastic-plastic interface where they are obviously violated. Approximate solutions, such as the one mentioned above, overlook the most important factor, namely the local plastic action, which strongly modifies the stress distribution. Errors by a factor of 2 should not be surprising. An additional frequent error is made in the problem of the symmetrical notched bar in plane strain. The value of the average stress at general yielding, is frequently found equal to the yield stress in simple tension, rather than about 2.5 times higher, as is known from limit analysis (1). This error has frequently been made, obviously taken over from the original work of Allen and Southwell (42) and Jacobs (43) who, however, followed much more accurate and thorough methods than the above mentioned. Their error was to assume that general yielding sets in the moment when the plastic zones from opposite symmetrical notches first touch each other, which does happen when the average net stress is not much greater than the yield stress in simple tension. It is now well known that this is not correct. Drucker (46) and Lee (47) have explained that the flow pattern needed to cause unrestricted plastic flow is not formed when the plastic zones first merge, but at a considerably higher average stress, as much as 2.57 times the yield stress for external deep parallel-edged notches.

Solutions based on an elastic-perfectly plastic material are of no great use in the problem of brittle fracture, even when they are correct. It is quite clear that in the range in which it may be idealized as perfectly plastic, steel does not fracture in a brittle manner. Since fracture does not occur below the general yield level of the idealized material, the elastic plastic solutions (contained plastic deformation) are irrelevant to brittle fracture. On the contrary, flow limit calculations, though based on an idealized perfectly plastic material, are relevant to undamaged ductile steel which can sustain large strains. The flow limit indicates when the strains become very large. At that stage strain hardening and constraints build up in the real material, and can raise the local stress to a very large value (of the order of the theoretical strength) at which it fractures. Perfectly plastic solutions are also inapplicable to fractures occurring below general yield (brittle), because these in-

a. Average net stress 90% of virgin yield

b. Average net stress 102% of virgin yield

FIG. 2. BIREFRINGENT COATING SHOWING LINES OF CONSTANT PRINCIPAL IN-PLANE SHEAR DIFFERENCES IN NOTCH REGION OF 3/4" THICK PLATES LEFT: AV. STRESS 28KSI. RIGHT: AV. STRESS 31.3 KSI.

dicate an embrittled material which cannot be approximated with a perfectly plastic law. In embrittled or low-ductility materials, strain hardening is rapid, so that the local stress can rise to the fracture value after little straining, as has been extensively discussed by Drucker (12). Such relatively small plastic strains are evident in all brittle failures. Parallel arguments employing a strain instead of a stress condition of fracture may also be used.

This discussion shows the futility of stress and strain calculations based on perfectly plastic materials, and even more so of approximate calculations. It also shows the importance of the determination of the general anisotropic stress-strain relations (tensorial relations) of work hardened steel. Unless this can be found or inferred from special tests, there is no clear way of calculating the stress or strain distribution around a notch and solving the problem of brittle fracture.

An experimental possibility of solving the problem would be to measure the strains at a notch. But this is extremely difficult, because interior and not merely the surface strains are needed. No direct means of interior stress measurement exists at present. Measurements of the strains in the midplane of an aluminum notched bar made up of two cemented halves marked with grids on their matching faces (51), have been made after ungluing. The best available synthetic cements, however, do not appear sufficiently strong to hold rigidly together work hardened steel plates deformed in the plastic zone. Another possibility would be to estimate the interior strains from the variation of the surface strains and from other measurements e.g. of surface curvature and thickness changes during the whole process of loading. For this purpose several tests were made with 10 in. square notched plates having birefringent coatings (52), attached to the metal surface. Figure 2 shows (4) such a plate with a 0.06 in. coating subjected to an average net stress of 90% of the yield point in the initial state (left) and 102% (right). The sensitivity is sufficient to show the individual Lüders' lines at low strains, which merge into a more uniform distribution at higher strains.

2. PRECOMPRESSED NOTCHED PLATE TESTS

The first object of the present work was to reproduce consistently in the laboratory true "brittle" failures of essentially unwelded steel plates, i.e., to obtain initiations of fracture under static loading at a low average net stress. The way to achieve this would be to exhaust the original ductility of the steel. To increase the chances of success, a 3/4 in. rimmed pedigree steel "E" of high brittle-transition range was chosen. Typical composition and properties of E-steel are given in Table I and in more detail in reference 38. Various methods of embrittling the steel were tried (2)-(3). The most efficacious was to prestrain machine-notched plates in the direction which would produce compression of the notch roots (3)-(4). Subsequently, the plates were tested in tension at a temperature of -5 to -14° F. The 10 in. square 3/4 in. thick plate had the sharpest possible milled 1 1/2 in. deep symmetric 90° notches on a pair of opposite sides and was subjected to compression on the other pair until 1.000 in. gage lengths across the notch roots shortened by amounts varying between 0.015 in. and 0.060 in. Clamping between heavier plates prevented buckling. The prestrained plate was next welded to special heads which would yield under tension so as to eliminate any eccentricity of loading. It was then covered with foamed plastic insulation, and with thermocouples fixed at several points was cooled to about -18 °F in a freezer. The loading was started at an initial rate of about 50,000 psi per minute. The rate slowed appreciably when the hinges started yielding.

The results were remarkable. Most prestrained plates fractured at a low net-stress level, the lowest at 12% of initial yield (3)-(4). As was found by about 100 tests (3),(4),(6), the general trend was toward lower stress fractures for higher prestrains. What is more important, however, is that the prestrained plates developed arrested cracks at extremely low net stress, usually between 9 and 30% of initial yield. Figure 1 shows such cracks on steel plates with ground faces which were not carried to ultimate failure. The brittleness at the point of initiation of the fracture is in clear contrast with the plastic deformation at the point of arrest. It should be noted that these cracks start under static conditions at low loads and at the ralative bluntness of the sharpest possible milled 90 deg. notch; that they are arrested at a greater depth and root sharpness; and that frequently they do not restart even at loads producing general yielding of the section. The result is totally contrary to the elementary energy theories of fracture of the Griffith type, which postulate an average stress at fracture inversely proportional to the square root of the crack length (14)-(16). The obvious

TABLE. I. TYPICAL COMPOSITION AND PROPERTIES OF STEELS.

Steel	Element, per cent									Yield Strength psi	Ultimate Tensile Strength psi	Elongation per cent		Charpy Impact	
	C	Mn	P	S	Si	Cu	Ni	Cr	Mo			In 8 in.	In 2 in.	ft-lb	Temp. deg. Fahr.
E	0.20	0.33	0.013	0.020	0.01	0.18	0.15	0.09	0.02	32 000	65 000	36	30	15 to 3.3	55 to -11
ABS-C	0.20	0.62	0.014	0.030	0.20	0.27				43 000	70 000	29			
A7	0.26	0.48	0.014	0.032						35 000	65 000	30			
T-1	0.12	0.69	0.011	0.030	0.17	0.31	0.88	0.56	0.44	111 000	120 000		20	15	-120
HY-80										80 000	95 000		24	140	-70

explanation of this behavior is the plastic compression of the steel in the region of the notch root and a consequent reduction of its ductility below the amount required when the flow limit is approached. The crack crosses the embrittled material and stops as it enters the more ductile region, unless it has picked up enough velocity to be able to propagate at the existing low stress level. Similar results have since been obtained with a British Admiralty steel (54), and with an ABS Class B steel.

Tests were also made with plates prestrained in tension and tested cold in tension in the direction of prestrain or transversely to it (2). The results were not as spectacular as with precompressed plates, but fractures in the parallel direction again occurred below the raised yield strength, and even below the initial yield point in the transverse direction. Cold extension appears to reduce the ductility anisotropically. Fracture anisotropy has also been found after hot rolling (35).

Unlike the earlier static test which produced fractures at the fixed yield stress, the present low average stress fractures permit the detection of the size effect as a change of the average fracture stress. Comparisons made to this effect between similarly compressed 10 and 20 in. plates (4), and between plates varying in width from 6.67 to 20 in. (7) did not disclose fracture stress variations consistent with a size effect.

3. REVERSED BEND TESTS

A series of tests with uniformly compressed bars was begun (section 7) in order to study the exhaustion of ductility by prestraining which led to the brittle fractures of the precompressed notched plates. These tests required a lengthy compression procedure with precautions against buckling, and great care in the machining of the tension specimens to prevent heating or straining. To get a rapid approximate answer on the effect of prestrain, Ludley and Drucker (8) devised the practical reversed bend test, shown in Figure 3, which has proved quite usefull (9)-(10). A similar test had been used by Lagasse and Hoffmans

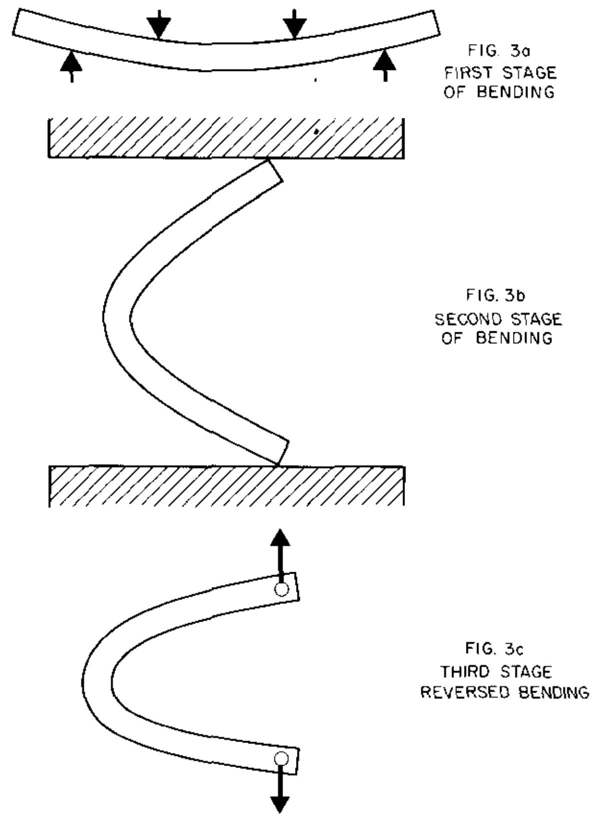


FIG. 3. REVERSED BEND TEST.

(27)-(29), in the study of metal forming, but involved a severe lateral pressure at the subsequently tested region of greatest interest. Although the strain varies with the depth from the free surface it is almost constant over an appreciable arc of the bent bar. The maximum prestrain is calculated from the radius of curvature and the thickness of the bar. The strain in a bent beam specimen is far more homogeneous than a notched specimen and far more reproducible. In fact the scatter of results was much smaller with the bent beam than with notched plate tests. The success of the reversed bent bar test raises the hope that it can provide an absolute type of test, if correlation is found with axially compressed bar tests and with field experience.

The bars used were 10 in. long and 0.75 x 1.00 in. in cross-section, where 0.75 in. was the as-rolled thickness of the parent plate. Tests with thinner or

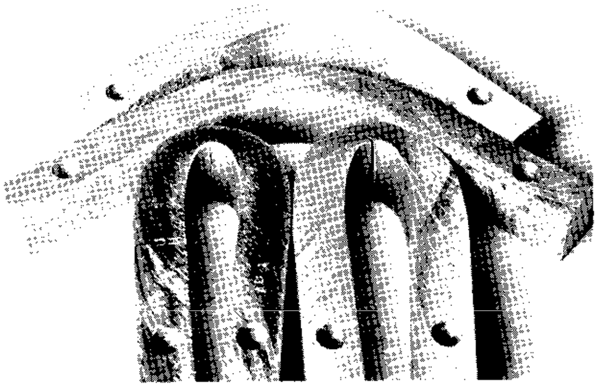


FIG. 4. BRITTLE AND DUCTILE BEHAVIOR OF 0.75 x 1.00 IN. BARS DURING REVERSED BENDING.

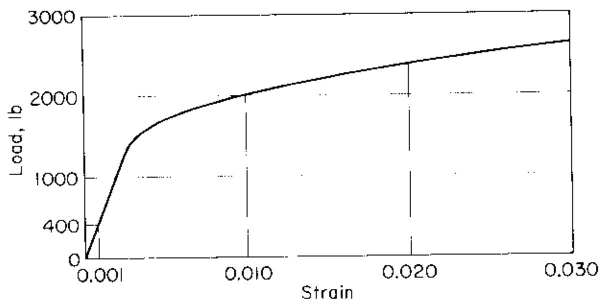


FIG. 5. SKETCH OF LOAD VS. MAX. STRAIN DURING UNBENDING OF BARS OF ABS CLASS C STEEL PRE-BENT TO 60% STRAIN.

wider bars failed to show any significant differences. The tests disclosed that the reduction of the initial ductility was not proportional to the compressive prestrain. The ductility appeared to remain consistently high until the prestrain reached a critical value, at which the ductility dropped suddenly to very small values. A possible reason for this was given by Drucker (12). The critical strain at which this change occurs is henceforth called the "exhaustion limit" and can usually be determined to within a strain of ± 0.02 or better. Figure 4 shows at the rear two bars of E-steel which were initially bent to strains of 0.57 (rearmost) and 0.30 (i.e. 57% and 30%), both below the exhaustion limit, and were ductile enough to be bent open to angles of about 120°, with strains of the order of 0.20 or more, under loads of more than 5000 lbs. In the front are two bars of an ABS Class C steel which were prestrained to strains of 0.63 (left) and 0.65 (right), which is just above the exhaustion limit for this steel. Both cracked and fractured at loads smaller than 2000 lb , with extensional strains of the order of 0.01. Photomicrographs of a section in the plane of bending of a bar of E-steel which fractured at very small load showed the compression of grains at the intrados and their extension at the extrados (9).

Obviously, the important characteristic is the prestrain causing the sudden drop in ductility, i.e. the exhaustion limit, and this can be easily determined by load measurements alone. A typical variation of strain at the intrados as a function of the applied load is shown in Figure 5. The bar was of ABS Class C steel precompressed to 0.60, (lower limit for fracture of unaged bars), and the measurement of strain during the final test was made with a strain gage cemented at the intrados. The rapid change of slope at about 1500 lb. corresponds to the onset of yielding. The continued positive slope is due to the spreading of the plastic zone toward the neutral axis, to strain hardening, and to the reduction of the moment arm by opening up of the bent bar. At a load of 5000 lb. the strain is estimated to be at least 0.10 and probably much larger (the strain gage was not working at such strain). In Figure 6 right, the fracture load of each bar is plotted against prestrain, but only up to 5000 lb., at which the tests were interrupted, as the strains were well above 0.10. This would happen with all bars prestrained to less than 0.50. On the contrary bars prestrained to more than 0.54 would all crack and fracture at loads of less than 1500 lb. The drop in ductility occurs at a prestrain of 0.52 ± 0.02. This sudden drop of load permits an arbitrary choice of a large load as limit between ductile and brittle behavior. The load of 5000 lb. seemed appropriate, but 4000 or even 3000 lb. would still give the same exhaustion limit.

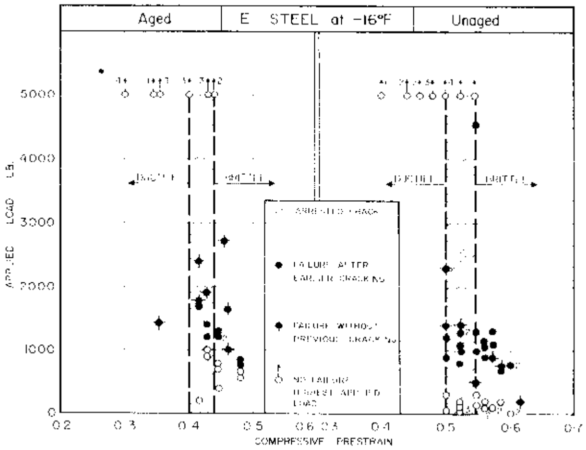


FIG. 6. RESULTS OF REVERSED BEND TESTS AT -16 F.

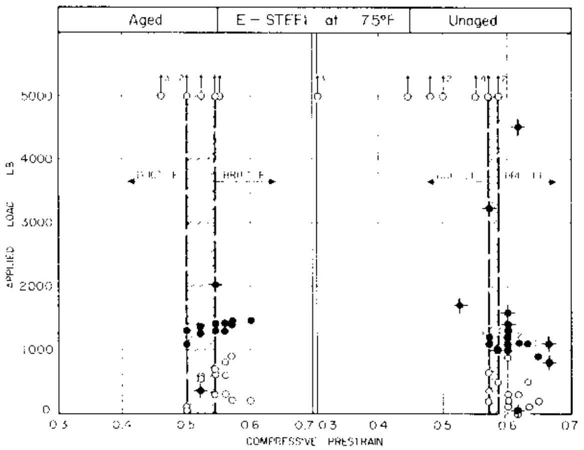


FIG. 7. RESULTS OF REVERSED BEND TESTS AT 75 F. (For Explanation of Point Symbols see FIG. 6)

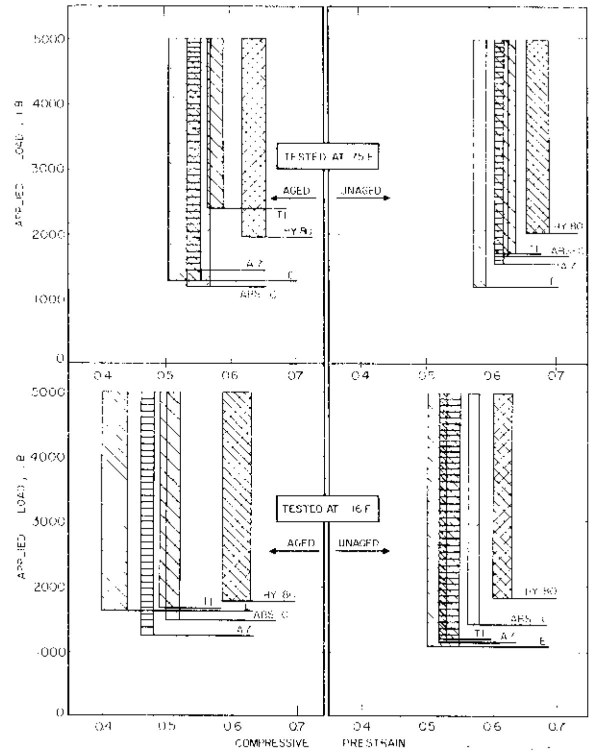


FIG. 8. COLLECTED RESULTS OF UNBENDING TESTS AT 75 AND -16 F OF STEELS E, A-7, ABS-C, HY-80 AND T-1.

It was found that both aging after straining (optimum accelerated aging consisted of heating for 1 1/2 hours at about 150°C), and lowering of the temperature of final testing reduced the exhaustion limit. This is indicated in Figures 6 and 7. A comparison (9) of E-steel with steels A-7, ABS Class C, T-1, and HY80 (properties in Table I) is given in Table II and Figure 8.

TABLE II. SUMMARIZED RESULTS OF REVERSED BEND TESTS.

Steel	Tested at -16°F		Tested at 75°F	
	Aged Exhaustion Limit	Unaged Exhaustion Limit	Aged Exhaustion Limit	Unaged Exhaustion Limit
E	0.40 to 0.44	0.53 to 0.55	0.50 to 0.55	0.57 to 0.59
ABS-C	0.50 to 0.5[illegible]	0.57 to 0.57	0.52 to 0.56	0.60 to 0.62
HY-80	0.59 to 0.63	0.60 to 0.63	0.61 to 0.65	0.65 to 0.69
A-7	0.46 to 0.48	0.52 to 0.55	0.52 to 0.55	0.61 to 0.62
T-1	0.49 to 0.52	0.52 to 0.53	0.56 to 0.59	0.60 to 0.64

TABLE III. REVERSED-BEND TESTS OF BARS WITH MACHINED SURFACES.
Size of specimen: 10" x 1" x 3/8"

Steel	COMPRESSIVE PRESTRAIN	APPLIED LOAD, LB.		STRESS $(4M/bd^2)$ksi	
		First Crack	Fracture	First Crack	Fracture
E	0.52	900	900	75.3	75.3
E	0.54	1 200	1 200	94.0	94.0
E	0.63	120	330	--	--
A7	0.50	--	++	--	++
A7	0.54	--	++	--	++
T1	0.50	--	1 800	--	++
T1	0.57	0	690	0	56.6
T1	0.60	0	690	0	58.7
E	0.60	240	310	20.8	26.8
A	0.60	420	500	35.8	42.5
A	0.63	120	420	10.4	28.7
T1	0.60	0	660	0	56.3
E	0.53	270	370	25.2	32.2
E	0.60	170	370	15.1	32.8
A7	0.52	--	++	--	++
A7	0.61	200	430	17.7	38.0
T1	0.54	0	750	0	64.0
T1	0.54	0	800	0	68.0

++ No fracture occurred up to a load of 2000 lb., corresponding to a stress larger than 90 ksi.

Tests were also made with beams of different widths, and with depths down to 1/4 in (8). Brittleness was induced at the same prestrain in all bars, which then fractured at similar stress levels, thus showing no size effect.

A limited number of tests were carried out with bars having the compressed face not in the as-rolled condition but machined to a depth of 3/8 in., leaving a 1 by 3/8 in. cross-section. Similar tests with bars machined down to 1/8. in. are reported in reference 8. Since no size effect exists, any variation in exhaustion limit should be attributed solely to the difference between as-rolled and machined surface. The results are given in Table III. The last two columns give the nominal stress at the first crack or at fracture for an assumed fully plastic stress distribution. This should be an acceptable approximation for the purpose of a comparison, when the loads and strains are large. The tests were not sufficient to determine sharp transition limits as was done previously. Wherever the results differ from those of bars of full plate thickness they appear to show a small rise of the exhaustion limit. Later tests with bars of ABS-C steel confirm that machined surfaces raise the exhaustion limit by about 0.05 (67).

The most remarkable result obtained from the reversed-bend tests is the sharp drop of the remaining extensional ductility at a certain value of the compressive prestrain. An interesting confirmation of this result is given in recent fatigue tests with precompressed specimens (34). The endurance limit was found to increase at small prestrains and to drop suddenly to low values at prestrains of about 0.50, which agrees well with the exhaustion limit determined by reverse-bend tests.

''Transition temperatures'' obtained with the material in its initial state have been extensively used in the assessment of steel's resistance to fracture. The present tests show that the important properties are those of the damaged steel, e.g. by prestrain and aging. Accordingly, the exhaustion limit and its dependence on temperature seem to be more significant than a transition temperature of the material in the initial state.

4. THE INFLUENCE OF RESIDUAL STRESSES IN THE NOTCHED-PLATE AND BEND-BAR TESTS

It is true that, besides an exhausted ductility, the notched plates and the bent bars had also a high residual tension at the region of fracture initiation. It has been argued that residual stresses and not the exhausted ductility may be the causes of fracture. Sound structures, however, are known to withstand successfully extremely high residual stresses. The subject of the role of residual stresses in fracture has been extensively discussed in recent years (55)-(64), and opinions have differed widely.

The influence of residual stresses can be studied in a rational way when not only stresses but the corresponding strains are considered (4), (11). Residual stresses are at most of yield or raised yield intensity and therefore may be wiped out or even transformed from tension to compression, by plastic strains of the order of the strain at the yield point (about 0.002). As is well known, plastic strains of the order of 0.010 or more are usually evident at the origin of even the most brittle fracture.

A detailed discussion of the importance of large fields of residual stresses (reaction stresses) and of the probable unimportance of localized stress has already been given (4) (11). The confusion over the role of residual stresses arises from their usual co-existence with prestraining and exhaustion of ductility. No clear decision can be reached on the relative importance of these two factors as long as they coexist. Clear differentiation can be done with tests involving each factor separately. Accordingly, the following 4 types of tests were conducted and the following answers were obtained:

a. Notched plates with residual stresses but *no prestrain* do not fracture in a brittle manner.

b. Bars uniformly precompressed to suitable prestrains, but free of initial stress, fracture at very small strains in typically brittle manner.

c. After removal of the residual tension without heating or plastic straining of the notches, precompressed notched plates still fracture at low average stress.

d. After removal of the residual tension without heating or plastic straining of the precompressed region, bars bent beyond the exhaustion limit still fracture in a brittle manner.

The answer is clear and unambiguous: In the present tests the localized residual stresses made no significant contribution to the initiation of brittle fracture. Initiation of fracture was caused by the exhaustion of ductility resulting from suitable prestraining and aging.

These conclusions, however, should not be interpreted as arguments against "stress-relieving". On the contrary, they emphasize the need for "stress-relieving" but indicate that its main function appears to be other than the removal of residual stresses. As will be shown in paragraph 6, stress-relieving causes a restoration of ductility. It should also be remembered that large fields of residual stress may have an influence on the initiation and more on the propagation of brittle fracture.

5. <u>REVERSED BENDING OF BARS STRAINED HOT</u>

The tests performed with precompressed notched plates, reversed-bent bars, and axially compressed bars (discussed in section 7) show that the exhaustion of ductility caused by suitable room temperature prestraining can lead to brittle fracture initiation under static loading alone. This is in agreement with observations on several service failures where the initiation was traced to cold-worked regions (65). The origin of brittle fracture, however, has also been traced to regions close to welds (65), though not at the welds themselves when they were sound. In effect this has led to various tests of plates containing a central longitudinal butt weld running over various types of notches in the welded edges (57)-(64). When cooled below zero, some of these plates develop arrested cracks originating at the welded-over notches, or fail at low-longitudinal loads. As indicated by Wells (13),(60), the zone adjacent to a weld is stretched by amounts up to 0.02 during cooling. Any notch or defect in this region will locally raise the strains by a substantial factor. The stretching caused by the shrinkage produces also large residual tension stresses which, unfortunately, have frequently been considered as the main cause of brittle fracture.

But, as was shown in section 5, the existence of strong local residual stresses does not cause brittle fracture if the steel has sufficient ductility. One is necessarily led to the conclusion that the steel has been embrittled at the root of the welded-over notches, and also at the corresponding points of fracture origin near welds in service failures. The heating due to

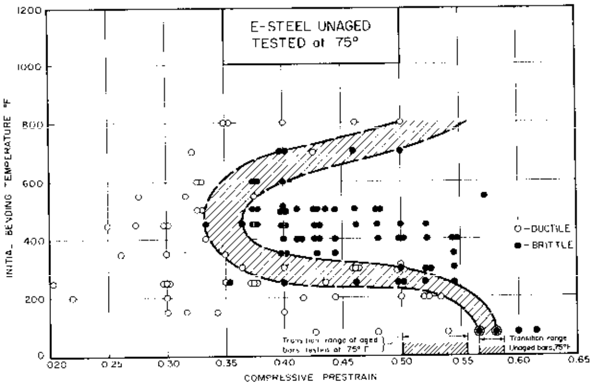


FIG. 9. REVERSED BEND TESTS OF UNAGED BARS OF E-STEEL PRESTRAINED AT VARIOUS TEMPERATURES AND TESTED AT 75 F.

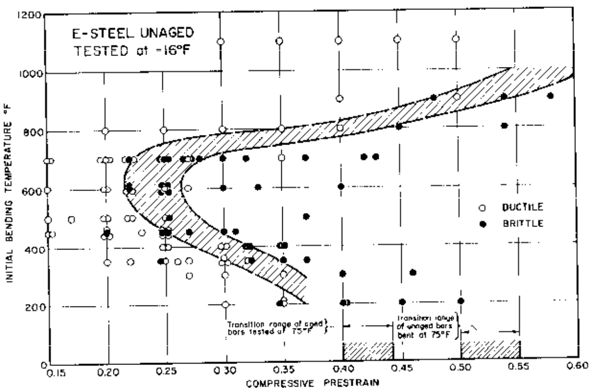


FIG. 10. REVERSED BEND TESTS OF BARS UNAGED E-STEEL PRESTRAINED AT VARIOUS TEMPERATURES AND TESTED AT -16 F.

welding, however, does not by itself cause such damage. On the other hand simple cold extension of notched plates (2), though causing some reduction of ductility, did not produce the extremely low load fractures achieved by precompression. It thus appears that the complicated hot prestraining occurring during the welding cycle should be the embrittling factor. The purpose of the hot-strained bar tests was to check the validity of this hypothesis (10).

As the actual straining at a defect close to a weld is very complex, changing from compression during heating to longitudinal and transverse stretching during cooling, it was decided at first to try only simple straining such as the compression or extension occurring during

TABLE IV. REVERSED BEND TESTS. INITIAL BENDING AT 150-250 F. TESTED AT 75 F. E -STEEL.

Bar No.	INITIAL BEND Temp. °F	INITIAL BEND Strain	APPLIED LOAD, LB. First Crack	APPLIED LOAD, LB. Fracture	STRESS ($4M/bd^2$) ksi First Crack	STRESS ($4M/bd^2$) ksi Fracture
1	150	0.32	--	+	--	+
2		0.34	--	+	--	+
3		0.40	--	+	--	+
4		0.46	--	+	--	+
5		0.48	--	+	--	+
6		0.50	--	+	--	-
7		0.52	--	+	25	+
8		0.52	--	+	--	+
9		0.52	--	+	--	+
10		0.54	--	+	--	+
11		0.56	100	1200	25	-
12		0.57	--	+	--	+
13		0.57	100	1300	25	-
14		0.63	100	1200	25	-
1	200	0.22	--	+	--	+
2		0.30	--	+	--	+
3		0.42	--	+	--	+
4		0.44	--	+	--	+
5		0.50	--	+	--	+
6		0.52	--	+	--	+
7		0.52	--	+	--	+
8		0.53	--	+	--	+
1	250	0.20	--	+	--	+
2		0.20	--	+	--	+
3		0.30	--	+	--	+
4		0.30	--	+	--	+
5		0.30	--	+	--	+
6		0.35	4200	4200	67	-
7		0.37	--	+	-	+
8		0.37	--	+	--	+
9		0.40	--	+	--	+
10		0.40	2400	2400	59	-
11		0.40	1300	1500	32	-
12		0.43	--	+	--	+
13		0.43	--	+	--	+
14		0.44	--	+	--	+
15		0.46	800	1100	19	-
16		0.50	4600	4600	--	-
17		0.54	300	1600	7	-

+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

TABLE V. REVERSED BEND TEST. INITIAL BENDING AT 300-400 F. TESTED AT 75 F. E STEEL.

Bar No.	INITIAL BEND Temp. °F	INITIAL BEND Strain	APPLIED LOAD, LB. First Crack	APPLIED LOAD, LB. Fracture	STRESS ($4M/bd^2$) ksi First Crack	STRESS ($4M/bd^2$) ksi Fracture
1	300	0.25	--	+	--	+
2		0.30	--	+	--	+
3		0.30	--	+	--	+
4		0.30	--	+	--	+
5		0.36	--	+	--	+
6		0.40	--	+	--	+
7		0.43	--	+	--	+
8		0.46	--	+	--	+
9		0.46	600	1300	15	+
10		0.46	--	+	--	+
11		0.46	--	+	--	+
12		0.50	100	1100	3	27.5
13		0.50	-	+	--	+
14		0.50	100	1300	3	-
15		0.52	100	1500	3	-
16		0.54	150	1300	3	-
1	350	0.20	--	+	--	+
2		0.21	--	+	--	+
3		0.30	--	+	--	+
4		0.35	--	+	--	+
5		0.40	600	1300	20	-
6		0.40	1600	3000	40	-
7		0.43	800	1200	20	-
8		0.44	50	900	1	-
9		0.54	200	1400	5	-
1	400	0.33	--	+	--	+
2		0.33	1100	1300	25	-
3		0.40	1300	1500	31	-
4		0.42	400	1100	9	-
5		0.43	600	1300	14	-
6		0.43	1350	1350	31	-
7		0.43	3700	3700	78	-
8		0.45	100	1600	3	-
9		0.46	700	900	16	-
10		0.48	900	1000	22	-
11		0.50	300	800	8	-
12		0.52	100	1300	3	-
13		0.54	100	1500	3	-
14		0.54	200	1100	5	-

+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

TABLE VI. REVERSED BEND TEST. INITIAL BENDING AT 450-600 F. TESTED AT 75 F. E-STEEL.

Bar No.	INITIAL BEND Temp. °F	Strain	APPLIED LOAD, LB. First Crack	Fracture	STRESS ($4M/bd^2$)ksi First Crack	Fracture
1	450	0.20	--	+	--	+
2		0.25	--	+	--	+
3		0.27	--	+	--	+
4		0.30	--	+	--	+
5		0.30	--	+	--	+
6		0.31	--	+	--	+
7		0.33	--	+		+
8		0.33	--	+	--	+
9		0.33	--	+	--	+
10		0.33	1000	1400	24	-
11		0.36	2200	2200	53	-
12		0.37	700	1300	17	-
13		0.37	900	2200	22	-
14		0.37	--	+	--	+
15		0.40	700	1300	17	-
16		0.43	1000	1600	24	-
17		0.40	700	1500	17	-
18		0.43	800	1200	19	-
19		0.46	700	1500	17	-
20		0.50	150	820	4	-
1	500	0.33	--	-	--	-
2		0.33	--	-	--	-
3		0.37	100	300	2	-
4		0.37	1300	1600	30	-
5		0.40	500	1500	12	-
6		0.40	300	1200	7	-
7		0.40	1600	1600	37	-
8		0.41	1000	1500	23	-
9		0.42	1900	1900	46	-
10		0.42	600	1800	14	-
11		0.44	100	1700	3	-
12		0.46	600	1500	14	-
13		0.46	800	1600	19	-
14		0.48	600	1900	14	-
1	600	0.33	--	+	--	+
2		0.37	50	1400	1	-
3		0.40	50	1200	1	-

+ No fracture occurred up to a load of 5000 lb. corresponding to a stress larger than 78 ksi.

TABLE VII. REVERSED BEND TEST. INITIAL BENDING AT 700-900 F. TESTED AT 75 F. E-STEEL.

Bar No.	INITIAL BEND Temp. °F	Strain	APPLIED LOAD, LB. First Crack	Fracture	STRESS ($4M/bd^2$)ksi First Crack	Fracture
1	700	0.32	--	-	--	+
2		0.33	2900	2900	64	-
3		0.35	2900	2900	64	-
4		0.37	2400	2400	57	-
5		0.40	2000	2200	48	
6		0.40	3600	3600	86	-
7		0.40	--	-	--	+
8		0.40	3100	3100	71	-
9		0.42	--	+	--	+
10		0.42	--	+	--	+
11		0.46	1400	1600	34	-
12		0.46	--	+	--	+
13		0.48	1700	1800	41	-
14		0.50	1300	1500	31	-
15		0.50	600	1500	14	-
16		0.50	--	+	--	+
17		0.51	1500	1800	36	-
18		0.52	--	+	--	+
1	800	0.35	--	+	--	+
2		0.40	--	+	--	+
3		0.46	--	+	--	+
4		0.50	--	+	--	+
5		0.57	--	+	--	+
6		0.57	--	+	--	+
7		0.60	100	1100	3	-
8		0.60	50	1200	1	-
1	900	0.42	--	+	--	+
2		0.48	2700	2700	65	-
3		0.50	--	+	--	+
4		0.54	--	+	--	+
5		0.58	400	1300	10	-

+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

TABLE VIII. REVERSED BEND TEST. INITIAL BENDING AT 200-400 F. TESTED AT -16 F. E-STEEL.

Bar No.	Temp. °F	INITIAL BEND Strain	APPLIED LOAD, LB. Arrested Crack	Fracture	STRESS ($4M/bd^2$)ksi Arrested Crack	Fracture	
1	200	0.30		+		+	Unaged
2		0.35		+		+	
3		0.35		+		+	
4		0.35	--	3300	--	+	
5		0.40	--	2000	--	50	
6		0.40	--	2700	--	68	
7		0.45	400	1400	10		
8		0.50	300; 700	1400	8; 18		
1	300	0.27		+		+	Aged
2		0.30		+		+	
3		0.35		+		+	
4		0.40	--	1700	--	43	
5		0.46	600	1300	15		
1	350	0.25	--	4850	--	+	Unaged
2		0.26		+		+	
3		0.30		+		+	
4		0.30		+		+	
5		0.32		+		-	
6		0.32	--	2000	--	50	
7		0.35	1300	1700	32		
8		0.37	--	3000	--	75	
9		0.20		+		+	Aged
10		0.22		-		+	
11		0.25		-		+	
12		0.27		+		+	
13		0.30		+		+	
1	400	0.25		+		+	Unaged
2		0.30		+		+	
3		0.32		1900		48	
4		0.35	1000	1500	25		
5		0.25		+		+	Aged
6		0.30		+		+	
7		0.32	1000	1500	25		
8		0.35	700	1400	18		

+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

TABLE IX. REVERSED BEND TESTS. INITIAL BENDING AT 450-600 F. TESTED AT -16 F. E-STEEL.

Bar No.	INITIAL BEND Temp. °F	Strain	APPLIED LOAD, LB. First Crack	Fracture	STRESS ($4M/bd^2$)ksi First Crack	Fracture	
1	450	0.15	--	+		+	Unaged
2		0.20	--	+		+	
3		0.20	2800	2800	70	+	
4		0.20	--	+		-	
5		0.23	--	-		-	
6		0.23	--	+		+	
7		0.25	--	+		+	
8		0.25	300	1000	8	-	
9		0.27	1200	1200	30	-	
10		0.30	100	600	3	-	
11		0.31	1100	1500	28	-	
12		0.10	--	+		-	Aged
13		0.15	--	+		-	
14		0.20	--	+		+	
15		0.25	--	+		+	
16		0.25	--	+		+	
17		0.25	--	3300	--	+	
1	500	0.15		+		+	Unaged
2		0.17		+		+	
3		0.20		+		+	
4		0.20		+		+	
5		0.22		+		+	
6		0.22		+		+	
7		0.25		-		+	
8		0.25	--	2600	--	65	
9		0.37	400	1400	10		
1	600	0.20		+		+	Unaged
2		0.22	--	2500	--	63	
3		0.22		+		+	
4		0.22	--	2900	--	72	
5		0.25	--	4400	--	+	
6		0.25	4700	4700	75		
7		0.25		+		+	
8		0.25	--	2900	--	72	
9		0.27		+		-	
10		0.30	--	1500	--	39	
11		0.33	--	1600	--	40	
12		0.40	150; 500; 600	1200	4; 13; 15		
13		0.15		+		+	Unaged
14		0.20		+		-	
15		0.22		+		+	
16		0.23	--	4100	--	+	

+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

TABLE X. REVERSED BEND TESTS. INITIAL BENDING AT 700-1100 F. TESTED AT -16 F. E-STEEL.

Bar No.	INITIAL BEND Temp. °F	INITIAL BEND Strain	APPLIED LOAD, LB. First Crack	APPLIED LOAD, LB. Fracture	STRESS ($4M/bd^2$) ksi. First Crack	STRESS ($4M/bd^2$) ksi. Fracture	
1	700	0.15	--	+		+	Unaged
2		0.20	--	+		+	
3		0.20	--	+		+	
4		0.22	--	4500	--	+	
5		0.22	--	+		+	
6		0.25	--	3600	--	+	
7		0.25	--	1900	--	48	
8		0.27	--	+		+	
9		0.27	--	3200	--	+	
10		0.27	--	5000	--	+	
11		0.30	--	3700	--	+	
12		0.32	1200	1400	30		
13		• 0.35	--	+		+	
14		0.37	2600	2600	65		
15		0.42	1200	1800	30		
16		0.43	--	2000	--	50	
17		0.15	--	+	--	+	Aged
18		0.15	--	+	--	+	
19		0.20	--	+	--	+	
20		0.22	--	+	--	+	
21		0.25	--	4900	--	+	
22		0.28	--	4100	--	+	
1	800	0.20	--	+	--	+	Unaged
2		0.25	--	+	--	+	
3		0.30	--	+	--	+	
4		0.35	--	+	--	+	
5		0.40	--	+	--	+	
6		0.45	--	2400	--	60	
7		0.54	1300	1500	33		
1	900	0.40	--	+	--	+	Unaged
2		0.48	--	2900	--	75	
3		0.50	--	+	--	+	
4		0.54	--	2800	--	70	
5		0.58	400	700	--		
1	1100	0.30	--	+	--	+	Unaged
2		0.35	--	+	--	+	
3		0.40	--	+	--	+	
4		0.45	--	+	--	+	
5		0.50	--	+	--	+	

\+ No fracture occurred up to a load of 5000 lb., corresponding to a stress larger than 78 ksi.

the initial bending of bars of the reversed bent test (section 3). The bars were heated at the required temperature for a period of 45 to 90 min., then given the initial four-point bending (Fig. 3a). After reheating for about 30 min., they were bent (Fig. 3b) to various radii and left to cool. About one day later they were finally tested in reversed bending (Fig. 3c) at 75°F, or at -16°F. The results of 122 tests at 75°F and of 84 tests at -16°F are shown in Figures 9 and 10 (10), and in Tables IV to X. The last two columns give the nominal stress at the first crack or at fracture for an assumed fully plastic stress distribution. The influence of the prestrain temperature is quite marked. Already at 200 to 250°F, the transition occurs at lower prestrains (the embrittlement is stronger) than at room temperature. The lowest transition limit occurs at the blue brittleness range around 450°F for reversed bending at 75°F, and around 600°F for reversed bending at -16°F. The transition prestrain for bending at 450 to 600°F is almost half as big as at room temperature. As expected, the transition prestrains are lower for -16°F than for 75°F. The lowest transition prestrain is only about 0.20, which is less than half the value found with bending at 75°F. The damage diminished when the temperature of prestraining was raised above 600°F as is obvious from the increasing transition prestrain. Above 900°F the effect of prestraining was smaller than at room temperature. Tests were made up to 1250°F where no reduction of ductility by prestraining could be detected. Similar tests were made with an ASTM A-7 steel. It was found again that precompression at about 600°F was far more damaging than at 75°F (10).

Tests in small numbers were also made to study the effect of hot extension on the cold ductility (10). For this purpose after initial hot bending and cooling to -16°F, the bars were subjected to a continued bending in the same manner as during initial bending (Fig. 3b). Because of shortage of E-steel, only bars of A-7 steel were used. It was found that embrittlement could be induced with extensional prestrains of the order of 0.35. In cold continued bending, these bars fractured at the extrados at additional extensions of 0.02 to 0.03. However, the scatter of the results was considerable.

The effect of hot straining had been studied by Körber, Eichinger, and Möller (21), and by several other investigators (35), (38)-(40). The embrittlement caused by relatively small strains at medium high temperature is of great significance because it appears to be the cause of the frequent service fracture initiations which are traced to defects close to welds.

6. RESTORATION OF DUCTILITY BY HEATING

6a. Purpose and Method of Testing. Work hardened steels, as e.g. by forming or spinning, are made again ductile by heating close to the annealing temperature. Lagasse and Hoffmans (30) have recently shown that considerable ductility may be restored after a heat treatment at about 1100°F. The present tests with bars of

E-steel are a first attempt to determine the relation between time and temperature needed to restore ductility. More detailed results obtained with controlled heat treatment of prestrained bars of ABS-B steel are reported separately (66.).

The previously developed method of controlled embrittlement and testing by reversed bending proved very useful in this investigation. Three steels were tested: E-steel, ABS Class C, and an A-7 steel (Table I). All three have been used extensively in earlier tests, and their exhaustion limits when prestrained at 75°F or 450°F, or tested at 75°F or -16°F have been reported (9)-(10). In the present tests the bars were prestrained beyond the exhaustion limits for the conditions under investigation, and were then heat treated at various temperatures between 1050° and 1500°F for various lengths of time. After cooling in air to 75°F, the specimens were tested in reversed bending in the usual way. Some specimens, however, were aged by heating to 300°F for 1 1/2 hours before heat treating. Their results are identical with those of bars which did not receive any separate aging treatment.

The heat treatment was done in an oven, which is not an ideal method for this purpose as the gradual heating obscured the exact time-temperature relationship which was sought. Nevertheless, with an identical temperature rise in all similar tests it would still be possible to get a close approximation of the time needed at each temperature for the restoration of sufficient ductility. In effect the time probably varies almost exponentially with the temperature, so that only a small temperature range close to the highest in each test should have a significant influence. Unfortunately the time spent in this range was unavoidably influenced by many factors such as position and number of bars in the oven, variations in circulation, etc. Thus the present results are only of a preliminary nature. In addition the number of tests was limited by the short supply of E-steel, but is sufficient to show the general trend of the heat treatment.

Typical heating curves of the bars are shown in Fig. 11 for oven temperatures

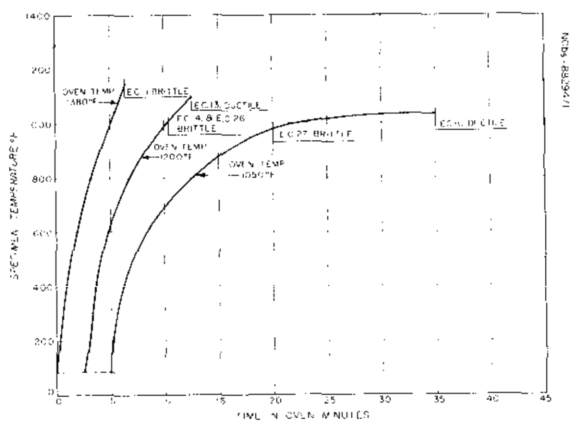


FIG. 11. TYPICAL HEATING CURVES.

of 1050, 1200, and 1380°F, as measured with thermocouples placed in holes drilled in the bars. In view of the uncertainty inherent to heating in an oven, no attempt was made in the present tests to determine the effect of a constant heat-treating temperature. Only the total time in the oven is reported for each bar. The temperature history may then be judged from the curves of Figure 11. Complete curves of heat treating time vs. prestrain for

TABLE XI. REVERSED-BEND TESTS OF BARS HEAT TREATED AFTER HOT INITIAL BENDING. E-STEEL.

Bar No.	HEAT TREATMENT Temp. °F	HEAT TREATMENT Time (min.)	INITIAL BENDING Temp. °F	INITIAL BENDING Compressive Prestrain	REVERSE BEND Temp. °F	REVERSE BEND Load (lb.) First Crack	REVERSE BEND Load (lb.) Fracture
EA3	850	5	450	0.46	75	100	1400
EC27	1050	15	450	0.44	75	2000	2000
EC15	1050	15	450	0.44	75	1300	1300
EC10	1050	30	450	0.44	75	--	+
EA17	1100	5	450	0.43	75	200	1400
EA19	1100	10	450	0.48	75	--	+
EA6	1100	15	450	0.46	75	--	+
EA5	1100	60	450	0.43	75	--	+
EA8	1160	12	450	0.43	75	--	+
E9	1160	19	450	0.43	75	--	+
E10	1160	21.5	450	0.43	75	--	+
E11	1160	25	450	0.43	75	--	+
E12	1160	29	450	0.43	75	--	+
EC14	1200	8	450	0.46	75	1400	1400
EC26	1200	8	450	0.48	75	1300	1300
EC13	1200	10	450	0.48	75	--	+
EC25	1225	15	450	0.50	75	--	+
EC12	1225	15	450	0.50	75	2200	2900
EC11	1375	6½	450	0.50	75	2100	2100

\+ No fracture up to a load of 5000 lb.

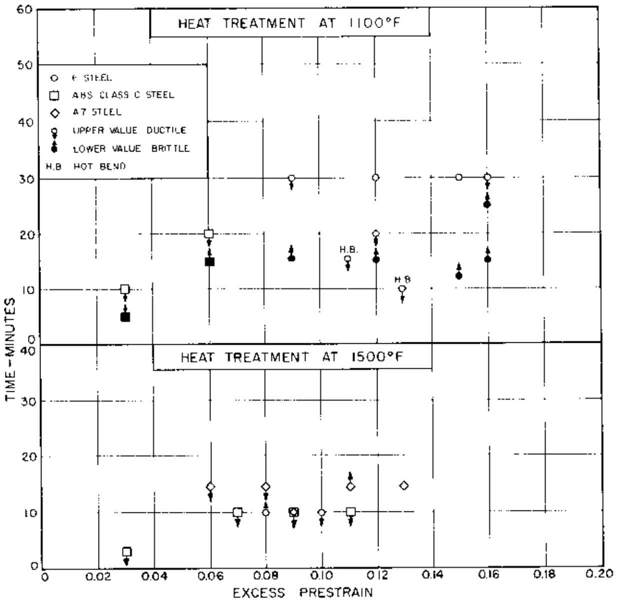


FIG. 12. EXPLORATORY HEAT TREATMENT OF PRESTRAINED STEEL.

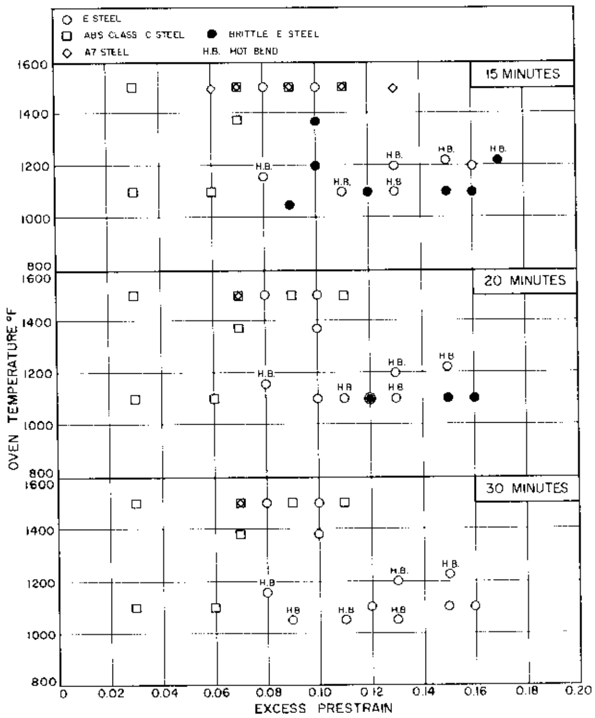


FIG. 13. EXPLORATORY HEAT TREATMENT OF PRESTRAINED STEEL.

various heat treating temperatures were determined in later extensive tests of ABS-B steel bars heated rapidly to the final temperature (60).

6b. Bars Prestrained Hot. As described in section 5, precompression by bending at about 450°F was found to be more damaging than at 75°F. The exhaustion limit for bending of E-steel at 450°F and reversed bending at 75°F was 0.35± 0.02 (Fig. 9). For the present tests the bars were bent at 450°F to prestrains in excess of the exhaustion limit by 0.05 to 0.15, so that without heat treatment they should all be brittle. The results are given in Table XI and are indicated by the letters H.B. in Figure 12 and 13. A period of ten to fifteen minutes in an oven heated to 1100°F restored the ductility of all hot prestrained bars. The amount of prestrain did not appear to influence the duration of the necessary heat treatment.

6c. Bars Prestrained at 75° F. The results obtained with E-steel prestrained at 75° F beyond its exhaustion limit, heat treated, and tested at -16°F and 75°F are given in Tables XII a,b. Similar results with ABS-C and A-7 steels are given in Tables XIII and XIV. Figure 13 shows the length of heat treatment at 1100° and 1500°F versus the excess prestrain (for aged specimens) of various bars, and indicates whether the bars were ductile or brittle. It appears that the minimum time needed to reductilize the bars at about 1100° increases with the excess prestrain. At 1500°F the necessary heat treatment was much shorter, down to 3 minutes in the oven for the smaller prestrains, and probably 10 for the higher, though some of the most severely prestrained bars remained brittle. It is obvious, however, that short heating times did not ensure either that the specimen had reached the oven temperature or that the temperature was uniform throughout the cross-section of the bars.

6d. Conclusions from Heat-Treating Tests. The present preliminary tests are insufficient for drawing exact conclusions, but they do indicate the following trends:

TABLE XIIa. REVERSED-BEND TESTS OF BARS HEAT TREATED AFTER INITIAL BENDING. E-STEEL.

Bar No.	HEAT TREATMENT Temp. °F	HEAT TREATMENT Time (min.)	INITIAL BENDING Temp. °F	INITIAL BENDING Compressive Prestrain	REVERSE BEND Temp. °F	REVERSE BEND Load (lb.) First Crack	REVERSE BEND Load (lb.) Fracture
E50	1100	3	75	0.54	-16	100	900
E57	1100	5	75	0.54	-16		
E49	1100	8	75	0.54	-16	600	600
E56	1100	10	75	0.54	-16	300	900
E62	1100	15	75	0.54	-16	200	900
E34	1100	20	75	0.54	-16	--	+
E23	1100	25	75	0.54	-16	--	+
E17	1100	30	75	0.54	-16	--	+
E31	1100	35	75	0.54	-16	--	+
E35	1100	40	75	0.54	16	--	+
63	1100	3	75	0.57	-16	100	900
55	1100	5	75	0.57	-16	100	1100
54	1100	5	75	0.57	-16	300	1300
64	1100	10	75	0.57	-16	300	1300
78	1100	12	75	0.57	-16	300	1600
45	1100	10	75	0.58	-16	50	1200
81	1200	3	75	0.57	-16	50	1100
79	1200	5	75	0.54	-16	300	1500
61	1200	8	75	0.54	-16	300	1050
59	1200	10	75	0.54	-16	200	1150
60	1200	12	75	0.58	-16	200	1050
84	1200	15	75	0.58	-16	700	1300
	1380	3	-16	0.52	-16	100	1100
	1380	5	-16	0.52	-16	100	1100
	1380	8	-16	0.52	-16		
	1380	15	-16	0.58	-16	700	1300
E19	1380	20	-16	0.52	-16	--	+
E37	1380	25	-16	0.52	-16	--	+
E2	1380	30	-16	0.52	-16	--	+

+ No fracture up to a load of 5000 lb.

TABLE XIIb. REVERSED-BEND TESTS OF BARS HEAT TREATED AFTER COLD INITIAL BENDING. E-STEEL.

Bar No.	Heat Treatment Temp. °F	Heat Treatment Time (min.)	Initial Bending Temp. °F	Initial Bending Compressive Prestrain	Reverse Bend Temp. °F	Reverse Bend Load (lb.) First Crack	Reverse Bend Load (lb.) Fracture
36	1100	20	75	0.54	-16	Brittle	Fracture
33	1100	25	75	0.54	-16	1700	1700
73	1100	30	75	0.54	-16	1000	1000
39	1100	30	75	0.54	-16	--	+
32	1100	45	75	0.54	-16	--	+
82	1100	20	75	0.57	-16	500	1300
27	1100	25	75	0.57	-16	600	700
43	1100	30	75	0.57	-16	2800	2800
22	1100	30	75	0.57	-16	--	+
39	1100	30	75	0.57	-16	--	+
40	1100	35	75	0.57	-16	--	4990
18	1100	35	75	0.57	-16	--	+
25	1100	40	75	0.57	-16	1920	2360
21	1100	45	75	0.57	-16	--	+
38	1100	45	75	0.57	-16	3100	3100
-	1100	45	75	0.57	-16	--	+
26	1100	70	75	0.57	-16	1830	3070
7	1100	100	75	0.57	-16	--	+
42	1100	20	75	0.58	-16	200	1000
68	1100	25	75	0.58	-16	500	800
44	1100	30	75	0.58	-16	--	+
41	1100	45	75	0.58	-16	--	+
83	1500	5	75	0.61	75	--	4180
69	1500	8	75	0.61	75	1270	2400
30	1500	10	75	0.61	75	1410	2370
5	1500	10	75	0.61	75	--	+
6	1500	10	75	0.63	75	--	+

+ No fracture up to a load of 5000 lb. or more

TABLE XIII. REVERSED BENDING TESTS OF BARS HEAT TREATED AFTER COLD INITIAL BENDING. ABS-C STEEL.

Bar No.	HEAT TREATMENT Temp. °F	HEAT TREATMENT Time (min.)	INITIAL BENDING Temp. °F	INITIAL BENDING Compressive Prestrain	REVERSE BEND Temp. °F	REVERSE BEND Load (lb.) First Crack	REVERSE BEND Load (lb.) Fracture
D12	1100	7	75	0.57	-16	500	1600
D11	1100	8	75	0.57	-16	300	1350
D74	1100	15	75	0.57	-16	3600	3600
19	1100	15	75	0.57	-16	--	+
15	1100	20	75	0.57	-16	--	+
13	1100	25	75	0.57	-16	--	+
2	1100	30	75	0.57	-16	--	+
D58	1100	3	75	0.58	-16	300	1400
D39	1100	5	75	0.58	-16	200	1450
D27	1100	8	75	0.58	-16	600	1000
D15	1100	15	75	0.57	75	--	+**
D65	1100	20	75	0.57	75	--	+
D7	1100	5	75	0.58	75	300	1350
D9	1100	10	75	0.58	75	--	4800**
D31	1380	5	75	0.61	75	400	1700
D69	1380	8	75	0.61	75	--	4500
D3	1380	10	75	0.61	75	--	4350*
D6	1380	12	75	0.61	75	--	4100**
D73	1380	15	75	0.61	75	--	+
D14	1380	20	75	0.61	75	--	+
D13	1500	3	75	0.57	75	--	+
D52	1500	5	75	0.57	75	--	4000**
D48	1500	8	75	0.57	75	--	+
D72	1500	10	75	0.57	75	--	+
D28	1500	12	75	0.57	75	--	+
D27	1500	15	75	0.57	75	--	+
D29	1500	20	75	0.57	75	--	+
D34	1500	25	75	0.57	75	--	+
5	1500	10	75	0.61	75	--	+
6	1500	10	75	0.63	75	--	+
D11	1500	15	75	0.65	75	--	4380**

\+ No fracture up to a load of 5000 lb. or more.

* Specimen behaved in ductile fashion initially, then failed in a brittle manner.

** Failure by gradual ductile tearing.

TABLE XIV. REVERSED BENDING TESTS OF BARS HEAT TREATED AFTER COLD INITIAL BENDING. A7-STEEL.

Bar No.	Heat Treatment Temp. °F	Heat Treatment Time (min.)	Initial Bending Temp. °F	Initial Bending Compressive Prestrain	Reverse Bend Temp. °F	Reverse Bend Load (lb.) First Crack	Reverse Bend Load (lb.) Fracture
12	1500	15	75	0.58	75	--	+
19	1500	15	75	0.60	75	--	+
--	1500	3	75	0.61	75	700	700
95	1500	5	75	0.61	75	870	870
135	1500	8	75	0.63	75	--	2730
--	1500	10	75	0.61	75	--	+
16	1500	15	75	0.63	75	1710	1710
96	1500	15	75	0.65	75	--	3870

\+ No fracture up to a load of 5000 lb. or more.

1. Heating to 1100°F for 30 minutes appears to restore sufficient ductility to all but the most highly strained bars.

2. The necessary duration of heat treatment shortens as the temperature rises above 1100°F. At 1500°F a few minutes appear to be sufficient for the medium and lower prestrains.

3. A few of the most severely prestrained bars (0.15 above exhaustion limit) remained brittle even after long heating at all the temperatures used. This may be a real irreparable embrittlement,

or may be caused by a deterioration of the steel by the long heating.

4. Bars of E-steel embrittled by hot straining need a much shorter heat treatment than bars strained cold.

5. A-7 steel appears to require longer heat treatment than E or ABS-C steels.

These tentative conclusions indicate that the so-called "high-temperature stress-relieving" treatment may be an efficient method for restoring ductility and preventing fractures, particularly in structures which have suffered embrittlement by hot straining close to welds. The heat treatment would be less efficacious with structures containing cold worked regions, particularly when the work hardening is very strong.

7. AXIALLY COMPRESSED BARS

7a. Method of Compression. Once systematic low static stress fractures in unwelded steel were achieved in precompressed notched plates, attention was focussed on the relation between ductility and compressive prestrain. This relation cannot be easily studied in the strongly variable strain distribution around the notches of the compressed plates, but requires a uniformly prestrained volume where both strain and stress may be directly measured. Such a uniform state of large deformation has already been achieved in a preliminary series of tests with axially compressed relatively slender bars laterally supported against buckling (5). It should be noted that the frictional constraint to lateral expansion of the bar ends may cause a local non-uniformity of straining which may extend over a length equal to the bar thickness. In relatively long bars this is of no consequence as only the uniformly strained middle portion is measured and tested. When the height is small in comparison with the bar thickness the two end regions merge and the strains are nowhere uniform nor accurately known. In effect steel cylinders sufficiently thick to avoid buckling were compressed at the National Physical Laboratory in England (31)-(32). At first they showed an ordinary barreling, but

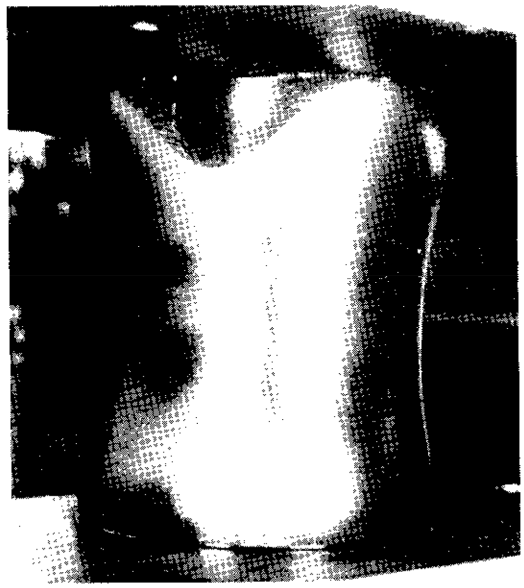


FIG. 14. PLASTIC CYLINDER SUPPORTING AXIALLY COMPRESSED BAR.

at higher prestrains deformed to a peculiar two-humped shape with a waist at mid-height. Considerable differences in strain could be inferred from polished and etched sections. The strain did show some uniformity over the central volume of the compressed cylinder, but varied considerably from the center to the circumference. The N.P.L. tests used specimens cut from the uniform central region, and were not concerned with the exact amount of the prestrain.

The buckling of axially compressed bars can also be prevented by using specimens with a reduced middle portion and guided longitudinal motion of the bar ends. Such specimens have also been used in tension (4). The strain distribution in the reduced portion of this specimen may be almost uniform in the elastic range. When yielding sets in, however, and especially at large strains, the constraining action of the thicker ends and the variation of thickness may cause a substantial variation of strain across the minimum section even when the change in sections is not very pronounced. Thus the overall change of diameter of the neck may not be an accurate measure of the maximum imposed strain.

In a first series of tests (5) cylindrical

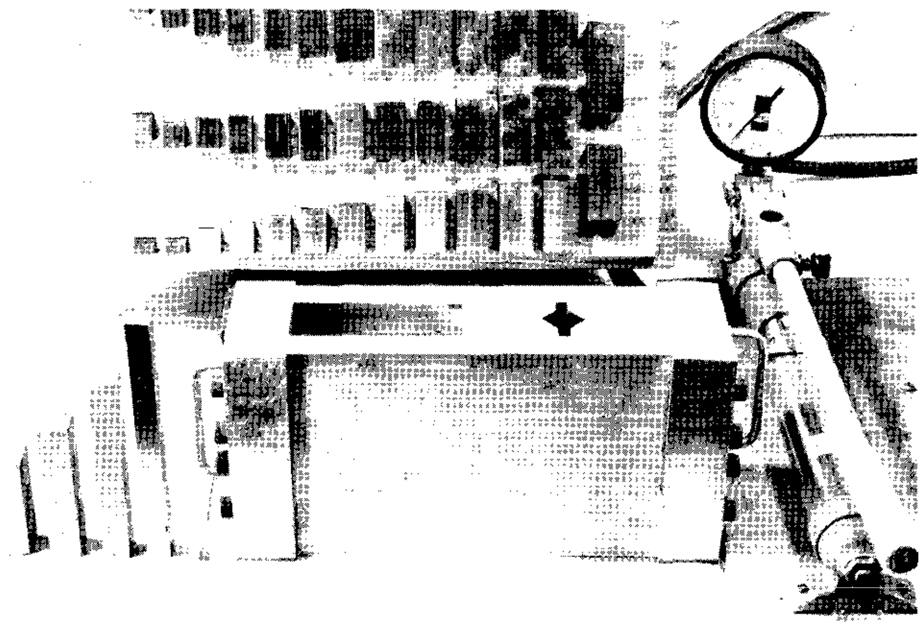


FIG. 15. COMPRESSION JIG WITH GUIDING V-BLOCKS, PLUGS, AND TEST BARS (LEFT) AT VARIOUS STAGES OF COMPRESSION.

bars of 0.75 in. diameter and 6 in. length were used. They were fitted in an axial hole bored in a plastic cylinder of 5 in. diameter. Bar and plastic cylinder were of equal length and were compressed together, so that the plastic buckling of the steel bar was prevented by the lateral elastic or viscoelastic support of the cylinder, Fig. 14. Compression was done in steps, large originally and smaller later, between which the bar was removed and the cylinder was machined to fit in the unstrained state the diameter and length of the compressed bar. The bar was also machined whenever the slightest non-uniformity was detected.

7b. The Compression Machine. The compression inside a plastic cylinder resulted in very lengthy delays for machining between compression steps and was abandoned in favor of a simpler method. Bars of a square cross-section were used and were laterally supported against buckling by two V-grooved blocks fitting over two diagonally opposite edges of the bar. This is shown in Fig. 15 where the V-blocks face each other and create a square hole. They are pressed on the specimen with a force of about 12000 lb , produced by a small horizontal hydraulic

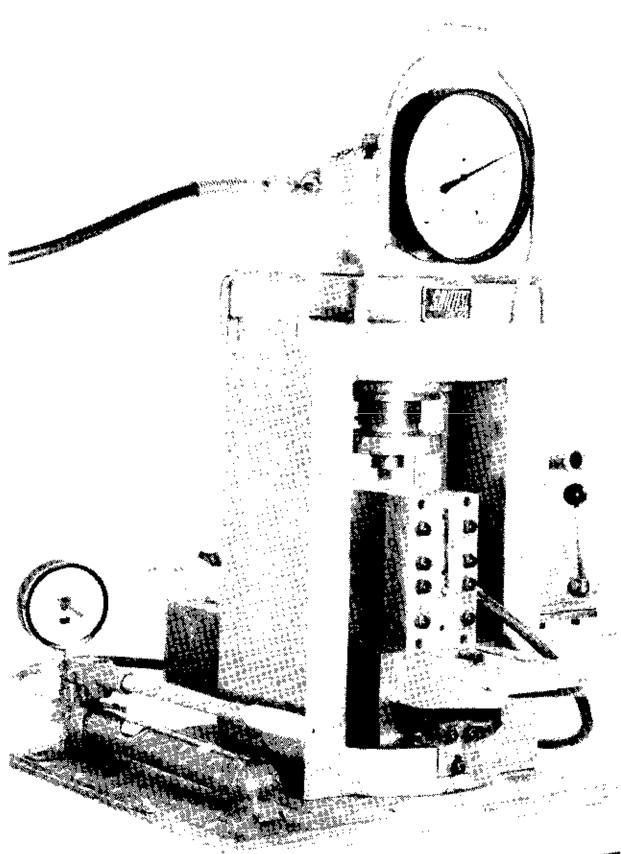


FIG. 16. COMPRESSION MACHINE.

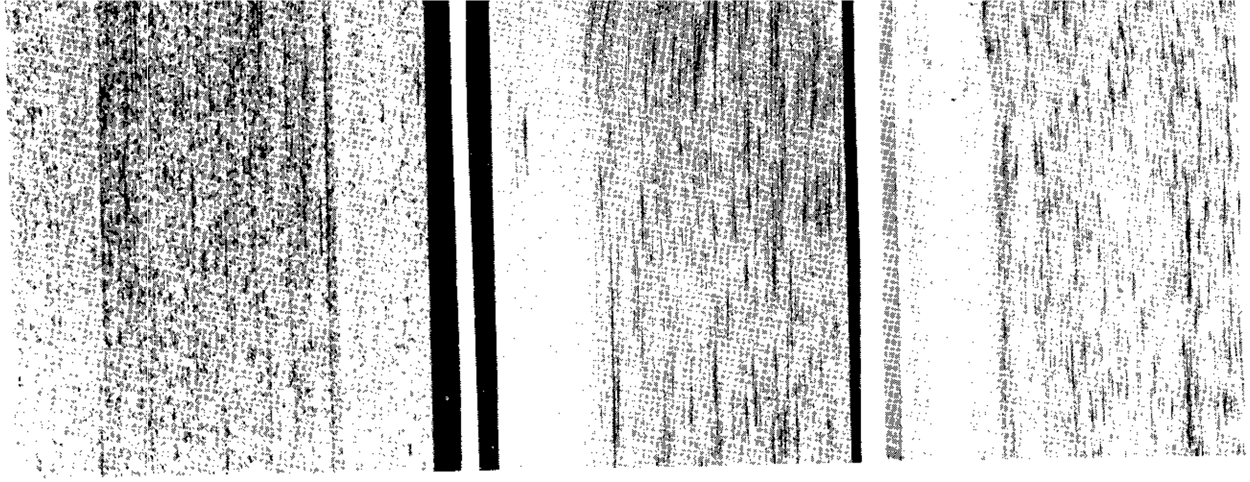


FIG. 17. ETCHED LONGITUDINAL SECTIONS OF BARS OF E-STEEL . a. VIRGIN BAR SHOWING SURFACE LAYERS WITH THINNER BANDING. b. LEFT HALF OF END PORTION AFTER 30% COMPRESSION. c. LEFT HALF OF END PORTION AFTER 50% COMPRESSION.

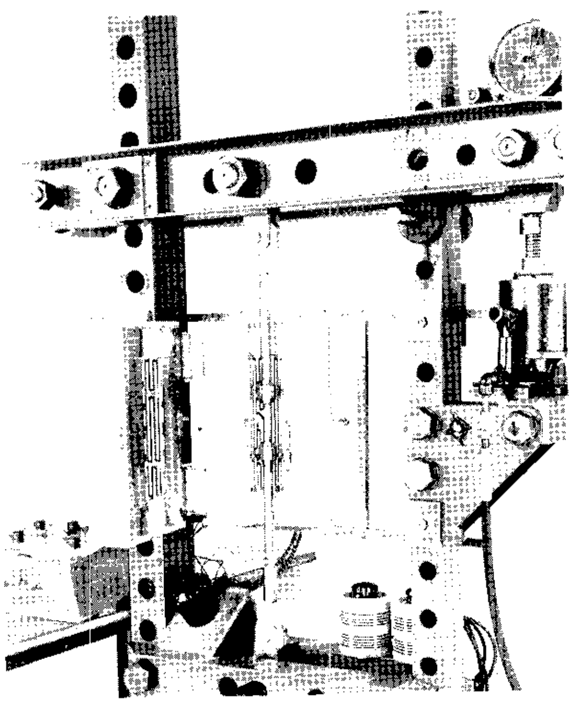


FIG. 18. AGING OF ROUND TEST SPECIMEN UNDER TENSION.

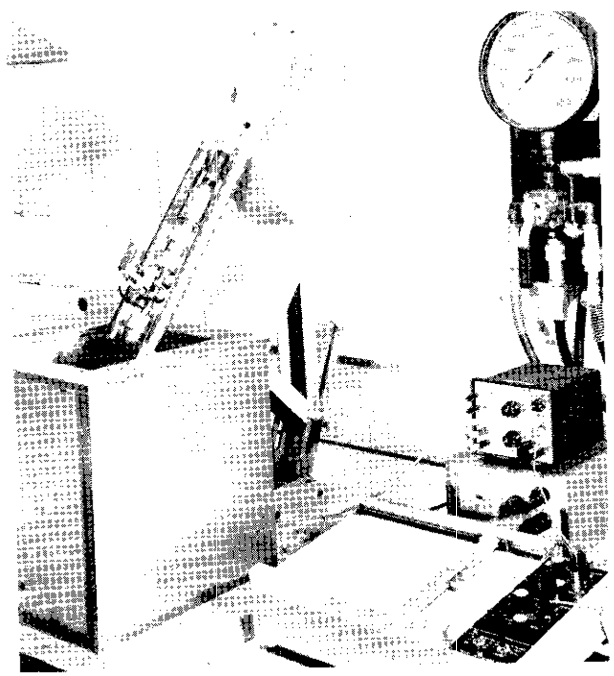


FIG. 19. SWINGING TEST MACHINE, TANK AND AUTOGRAPHIC EQUIPMENT FOR TENSION TESTS AT -16 F.

cylinder placed in the cavity at the left of the V-blocks and operated with the hand pump shown on the right. The whole compression jig is placed in a 200,000 lb. compression machine, as shown in Fig. 16. Figure 15 shows (background) the array of hardened plugs used to compress the bars, and on the left four bars at various stages of compression from an unstrained 9 in. long bar down to 58% compression. Friction between bar and V-blocks, and between bar ends and compression heads is reduced with the help of a 0.001 in. thick Teflon sheet. This reduces appreciably the axial load needed to produce the same strain, and prevents the uneven lateral expansion (thicker at top) of the bars.

The uniformity of straining of the bars was checked on longitudinal sections polished and etched (50% hydrochloric acid at 80°C) to show flow lines. Figure 17a shows a section across the thickness of an unstrained 3/4 in. plate of E-steel, in the rolling direction. Figures 17b and c show sections across bars prestrained by 30% and 50% respectively. About one third of the width of the bars has been removed after prestraining (to the right of Figures 17b and c). All three photographs show a distinctly finer banded structure in two outer layers and a coarser at the interior. The bands remain parallel even at the higher prestrain, except at a small region at the ends of the bars, and within a distance of about half the bar thickness from the ends. No macroscopic variation of strain across the bar thickness is evident.

The compressed bars were stored in a freezer at -18°F to prevent unwanted aging, except for the few hours required for machining into 0.505 in. tension specimens. The machining was done under a coolant and with great care to avoid heating or straining. Standard 3/4"-14 threaded heads were first made on all bars prestrained below 50%, but were increased to 1"-10 at higher prestrains, when some 58% tension bars broke at the threads.

7c. <u>Aging Under Tension</u>. Tests were made with unaged bars as well as with bars aged without load or under tension up to 75,000 psi. Aging under tension was done in a special straining frame with a cylindrical three-zone oven and a hydraulic-pneumatic loading system which kept the load constant over long periods of time. Figure 18 shows the straining frame, oven, and specimen in its holders.

7d. <u>Tension Machine with Cooling Bath.</u>

Tension tests were made at 75° F and at -16°F. Almost all 75°F tests were done in an ordinary testing machine, at first with a dial extensometer and later with an autographic extensometer, up to strains of a few thousandths so as to determine the 0.1% offset yield strength. Higher strains were calculated from the change of diameter which was measured with a dial gage. The bars to be tested at -16°F, however, would warm up before the test was finished. This was corrected with a specially constructed miniature testing frame which had narrowly spaced slender columns and could swing and be immersed into a cold tank bath at -16°F. Loading was done with a 21,000 lb. low friction hydraulic cylinder and pulling heads with spherical self-aligning seats. Figure 19 shows the testing frame swung out of the tank, with a specimen and extensometer held in place. A strain gage load cell in series with the specimen and a linear variable differential transformer type extensometer with associated demodulator provided the signals for plotting the load-elongation curve on an x-y recorder (right foreground). For each test the frame with mounted specimen was swung out of the coolant, the extensometer was attached and loading with autographic recording up to strains of about 0.010 was rapidly completed (less than 1 min.). The extensometer was then removed without interruption of the loading, the frame was swung into the bath and the testing was continued to fracture. No diameter measurement was taken during the test.

Use of the coolant was beneficial also for reducing the heating due to plastic deformation of the specimens which did not fracture at low strain. For this reason a pure water bath was used in the later tests at 75°F.

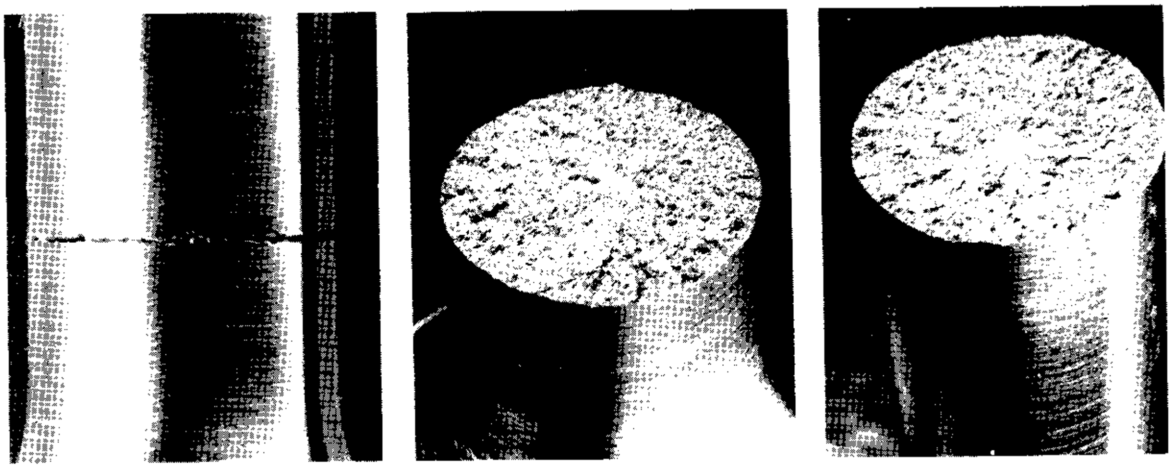


FIG. 20. FRACTURE OF BAR E-146: $\epsilon_o = 0.61$, $\epsilon_f = 0.023$.

FIG. 21. BAR E-114
$\epsilon_o = 0.61$, $\epsilon_f = 0.021$.

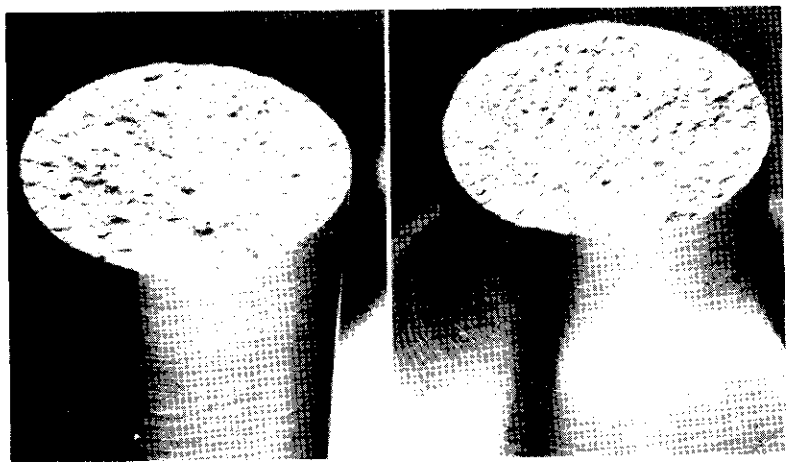


FIG. 22. FRACTURE OF BAR E- 137a
$\epsilon_o = 0.58$, $\epsilon_f = 0.04$.

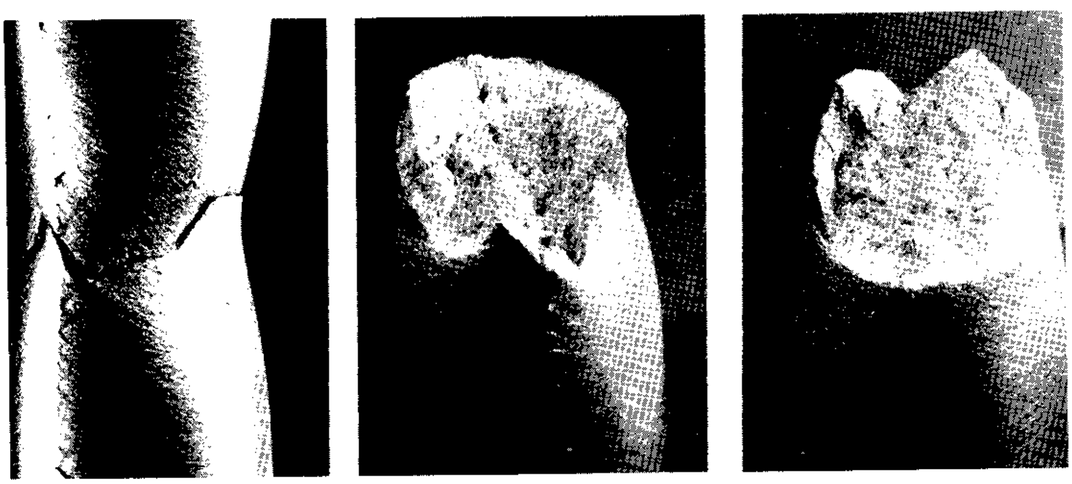


FIG. 23. FRACTURE OF BAR E - 157 $\epsilon_o = 0.61$, $\epsilon_f = 0.60$.

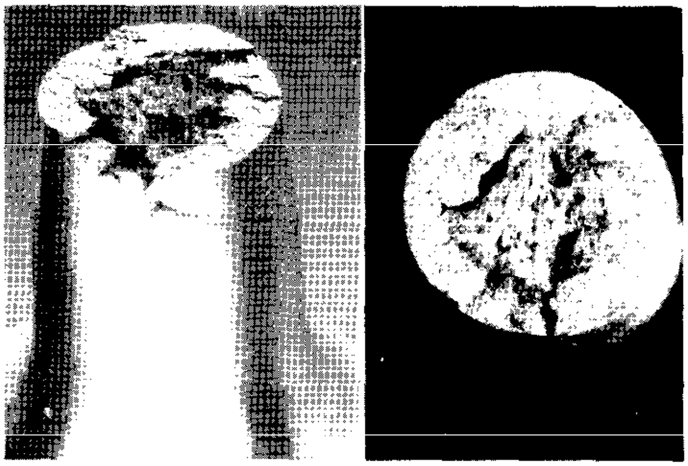


FIG. 24. FRACTURE OF BAR E - 111
$\epsilon_o = 0.67$, $\epsilon_f = 0.40$.

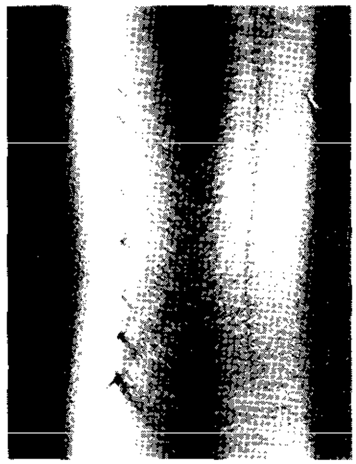


FIG. 25. BAR E- 117
$\epsilon_o = 0.55$, $\epsilon_f > 0.37$.

It is true that water as well as glycerine may affect the fracture strength. Water, however, is unavoidable, as it condenses and freezes on the cold specimens, but it is acceptable as reproducing standard service conditions. The corrosive action of glycerine on steel is not exactly known, but even if it differs appreciably from water, the effect should be altogether negligible in the short duration of a tension test. Even with an energy approach to fracture, the reduction of the surface energy due to either liquid should not be important. Even in the most brittle fracture the gross strains are of the order of 0.02 and the local strains are considerably higher. Surface energy is negligible in comparison with plastic strain energy. In addition, tests with and without a bath did not indicate differences which could be attributed to the surface effects of the coolant.

7e. Brittle and Ductile Behavior. Brittle fractures were achieved at the highest prestrains. Figure 20 shows a typically brittle fracture (prestrain $\epsilon_o = 0.61$;strain at fracture $\epsilon_f = 0.023$) which may have started at the center. Figures 21 and 22 show fractures which appear to have started at or close to the surface (prestrain 0.61 and 0.58; strain at fracture 0.21 and 0.04 respectively).

Ductile fractures showed an irregular, jagged cup-and-cone surface as in Figure 23 ($\epsilon_o = 0.61$; $\epsilon_f = 0.60$). Sometimes bars fracturing at large strains (order of 0.50, with pronounced neck) showed a central irregular region with some fibrous appearance, surrounded by a flatter region with steps, similar in texture with the most brittle fracture surfaces, as in Figure 24 ($\epsilon_o = 0.67$; $\epsilon_f = 0.40$). This pronounced difference from center to circumference raised the question of the uniformity of prestraining, and led to the macroetch studies (Figure 17), which indicated a high uniformity of macroscopic strain.

An interesting phenomenon appeared on the surface of ductile bars after the neck became evident, but well before fracture. A series of pinhole cavities with yield bands emanating from them at about 45° to the bar axis appeared and multiplied mostly along the same generator (Figure 25; $\epsilon_o = 0.55$; $\epsilon_f = 0.37$) and grew with the load. On one occasion the generator with the pinholes was identified with the trace on the specimen surface of a plane in the banded structure (perpendicular to the planes of the macroetch surfaces of Figure 17). The bands at 45° do not appear to be intersecting slip zones causing the cavity, but rather yield zones caused by and starting at the circumferential extremities of the cavities. This conclusion is based on the observation that the parallel zones from opposite sides of a cavity are not aligned segments forming a shape like the letter X, but are offset by a length of the order of the width of the

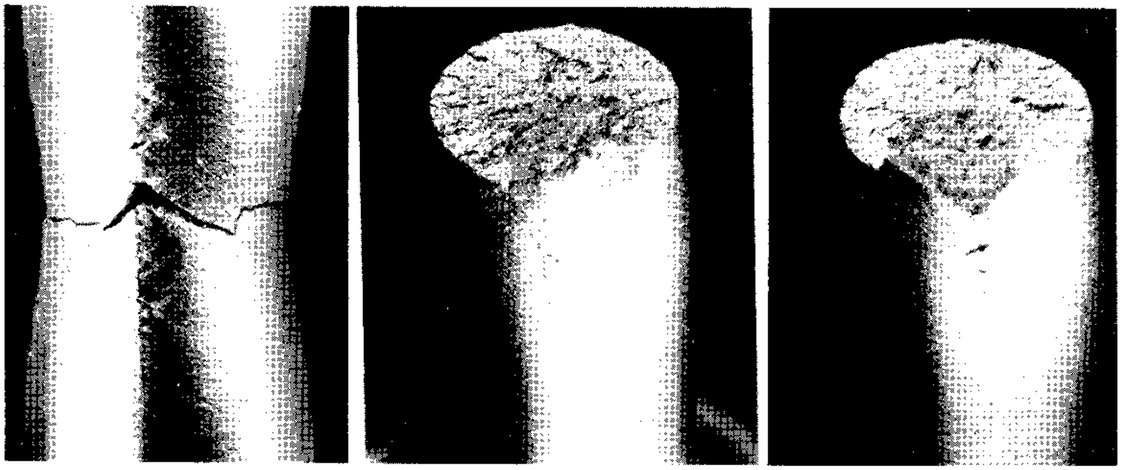


FIG. 26. FRACTURE OF BAR E - 139 $\epsilon_o = 0.58$, $\epsilon_f = 0.46$.

cavity as in the form > < . The fracture surface usually contained one such cavity, which frequently appeared to be the surface trace of an inner earlier fracture, as is clearly shown in Figures 24 and 26 ($\epsilon_o = 0.67$; $\epsilon_f = 0.40$). These interior fractures appear to produce the central region of different texture observed in some fractured bars, and also to be the cause of necking.

At the present no direct relation is found between these pinhole cracks or interior fractures which seem to occur at strains of the order 0.10 or more, and brittle fractures which occur at strains of the order or 0.01. It is surprising, however, how a series of local fractures, as indicated by the pinholes of Figure 25, not only fail to initiate a complete fracture in a severely work hardened bar, but even allow it to sustain an increased stress. Considerable ductility is present even after compressive prestrains as large as about 0.60, as is also evident from the large overall straining at fracture of such bars (0.60 to 0.80 or more).

7f. Test Results. The test results are given in detail in Tables XV a-d for bars tested at 75°F, and XVI a-c at -16°. They are tabulated according to increasing compressive prestrain, and show the 0.1% offset, the natural strain at fracture (equal to $\log_e r$ where r is the ratio of initial to instantaneous area at fracture), the true stress at fracture (based on the instantaneous fracture area), and the conditions of aging.

The natural strain at fracture plotted against the prestrain is shown collectively for all bars tested at 75°F in Figure 27, and separately according to whether they are unaged; aged without load; aged under 15 to 50 ksi; or under 60 to 75 ksi, in Figures 28 and 29. Likewise collected results of all bars tested in tension at -16°F are shown in Figure 30, and according to aging group in Figures 31 and 32.

These results fully confirm the drop of extensional ductility at a relatively narrow range of compressive prestrains, and the amazingly small effect of prestrains of lower magnitude. The scatter of the results is greater than with the reversed-bend bars and does not permit as sharp a definition of the exhaustion limit (with reversed-bend bars this was possible to within strains of ± 0.02). It is not possible to distinguish clearly between the effects of different aging procedures of bars tested at 75°. With bars tested at -16°F the exhaustion limit appears to decrease from about 0.61 to 0.50 as the aging load increases. The effect of a change of temperature from 75° to -16°F is more distinct. With few exceptions the exhaustion limit at 75°F lies between about 0.59 and 0.66, and at -16°F between 0.50 and 0.61. These are larger by 0.06 to 0.10 than the limits found with reversed-bend bars

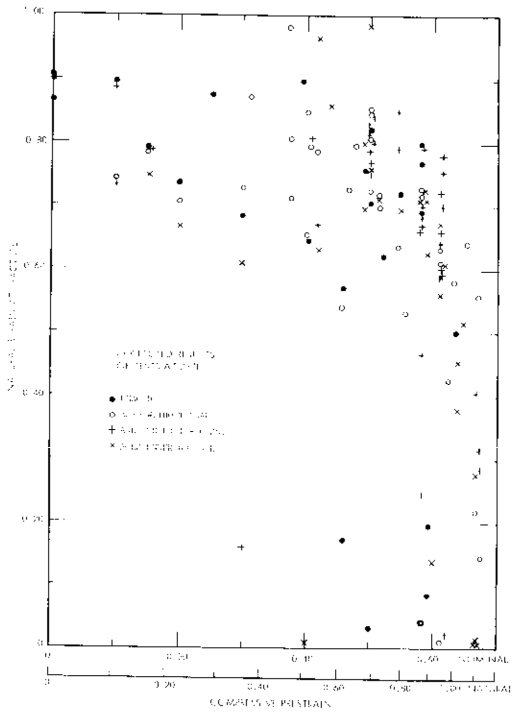


FIG. 27. COLLECTED RESULTS OF PRE-STRAINED BARS OF E-STEEL TESTED AT 75 F.

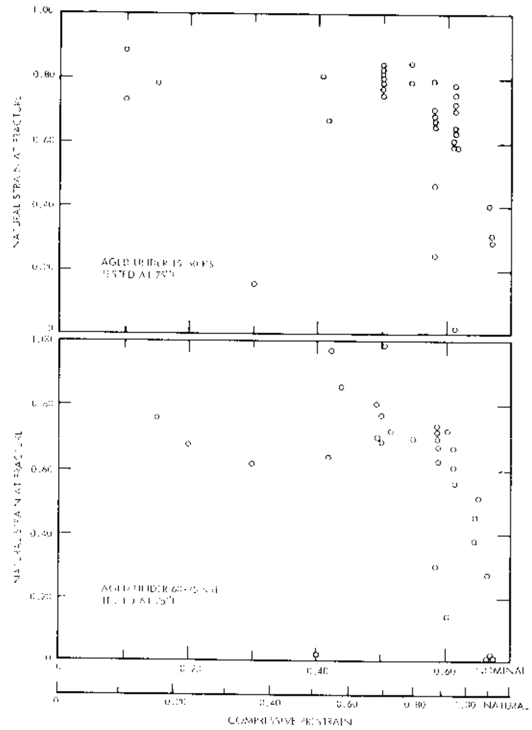


FIG. 29. NATURAL STRAIN AT FRACTURE VS. PRESTRAIN OF BARS OF E-STEEL, AGED UNDER 15-50 AND 60-75 KSI AND TESTED AT 75 F.

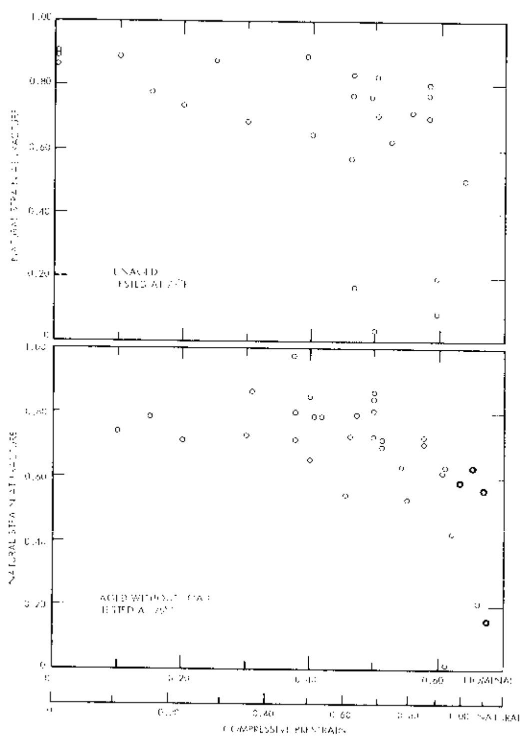


FIG. 28. NATURAL STRAIN AT FRACTURE VS. PRESTRAIN OF BARS OF E-STEEL, UNAGED AND AGED WITHOUT LOAD AND TESTED AT 75 F.

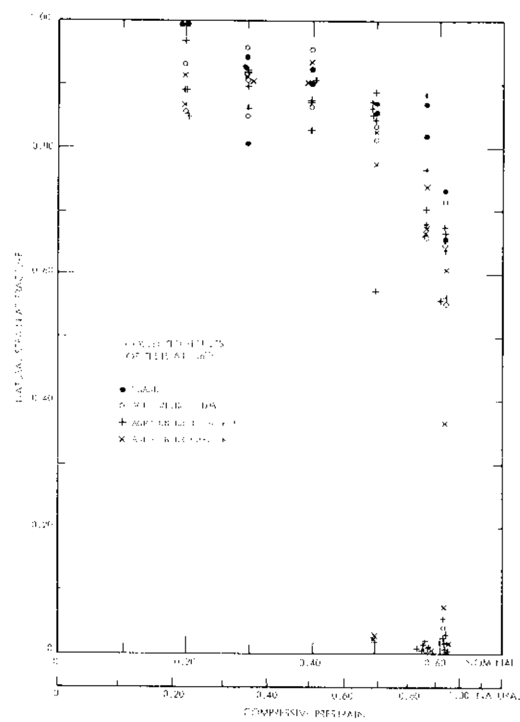


FIG. 30. COLLECTED RESULTS OF PRE-STRAINED BARS OF E-STEEL TESTED AT -16 F.

TABLE XVa. BARS AXIALLY COMPRESSED AT 75 F. TENSION TESTS AT 75 F. E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension during Aging ksi
[illegible]		5	0.60	119	Unaged
47	[illegible]	28	0.90	1[illegible]	Unaged
65	[illegible]	9	0.37	116	Unaged
4	0.14	--	0.70	112	Unaged
7	0.10	--	[illegible]	110	0[a]
11	0.11	--	0.64	115	[illegible]
13	0.12	--	0.74	109	50
8	0.13	--	0.79	1[illegible]	Unaged
6	0.13	--	0.79	109	[illegible]
10	0.15	--	0.74	101	[illegible]
11	0.15	--	0.79	110	[illegible]
5	0.22	--	0.74	106	Unaged
16	0.[illegible]	--	0.71	107	
9	0.20	--	[illegible]	[illegible]	43
3	0.[illegible]	--	0.65	115	Unaged
18	0.30	--	0.69	101	Unaged
12	0.30	--	0.79	113	0
15	0.30	--	0.16	55	46
16	0.31	--	0.91	110	75
39	0.31	56	0.87	131	0
35	0.37	57	0.98	113	[illegible]
16	0.38	[illegible]	0.81	13[illegible]	0
76	0.38	78	0.71	135	[illegible]
37	0.39	36	[illegible]	135	Unaged
17	0.40	--	0.[illegible]	105	Unaged
20	0.40	--	0.63	111	0
21	0.40	58	0.55	121	[illegible]
1	0.40	--	0.32	85	46
3	0.41	65	0.79	131	[illegible]
5	0.41	71	0.81	1[illegible]	45
17	0.41	16	0.79	133	0
11A	0.[illegible]	74	0.67	126	30
1	0.42	78	0.63	173	75
31	0.42	72	0.96	[illegible]	0 - 75
41	0.44	75	0.54	136	0 - 75[b]
21	0.46	--	0.37	110	Unaged
19	0.46	--	0.42	121	0
13	0.[illegible]	--	0.76	94	Unaged
14	0.47	36	0.[illegible]	113	0
38	0.56	60	0.84	117	[illegible]

a 0 denotes aged without load.
b Aged after each step of 5% compression.
c Aged under a continuously rising tension from 0 to 75 ksi.

TABLE XVb. BARS AXIALLY COMPRESSED AT 75 F. TENSION TESTS AT 75 F. E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension during Aging ksi
5A	0.49	46	0.76	1[illegible]	Unaged
55	0.49	62	0.86	134	60
43	0.49	57	0.86	1[illegible]	75
X-203	0.50	55	0.71	127	Unaged
151	0.50	52	0.81	134	Unaged
177	0.50	60	0.03	59	Unaged
11A	0.50	56	0.[illegible]	13[illegible]	0[b]
175	0.50	63	0.85	138	3[illegible]
150	0.50	64	0.81	135	0
212	0.50	59	0.85		
176	0.50	61	0.79	1[illegible]	15/15[d]
189	0.50	65	0.51	13	15/15
172	0.50	67	0.6[illegible]	133	27/30
174	0.50	66	0.73	136	27/30
149	0.50	6[illegible]	0.81	135	27/40
158	0.50	67	0.83	1[illegible]	27/40
152	0.50	65	0.77	135	27/50
134	0.50	70	0.75	1[illegible]	27/60
148	0.51	80	1.00	157	27/70
6	0.54	79	0.85	131	0 - 75
20	0.51	--	0.8	137	Unaged
8	0.52	37	0.74	125	0
4	0.52	[illegible]	0.[illegible]	1[illegible]	[illegible]
7	0.5	72	0.77	126	0 - 75[c]
18	0.55	--	0.73	141	Unaged
108	0.55	60	0.61	119	0
111	0.55	67	0.85	113	15/30
116	0.55	73	0.79	149	18/34
117	0.55	71	0.57	120	27/40
119	0.55	50	0.70	126	27/60
53	0.56	--	0.55	119	0
164	0.58	61	0.77	133	Unaged
180	0.58	6	0.61	1[illegible]	Unaged
Y-207	0.58	52	0.69	132	Unaged
136	0.58	65	0.73	1[illegible]	[illegible]
169	0.58	78	0.72	1[illegible]	[illegible]
1[illegible]	0.58	69	0.70	134	15/15
15[illegible]	0.51	6[illegible]	0.67	126	15/15

a 0 denotes aged without load.
b Aged after each step of 5% compression.
c Aged under a continuously rising tension from 0 to 75 ksi
d First number denotes stress during first 1/2 hour when temperature rises, second number the stress at aging temperature.

TABLE XVc. BARS AXIALLY COMPRESSED AT 75 F. TENSION TESTS AT 75 F. E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension During Aging ksi
15[illegible]	0.5[illegible]	71	0.95	93	27/[illegible]
187	[illegible]	[illegible]	[illegible]	118	27/[illegible]
175	0.56	--	0.79	136	27/30
139	0.56	73	0.47	1[illegible]	27/30
184	0.58	Flaw	Flaw	Flaw	27/4[illegible]
185	0.58	7[illegible]	0.66	118	27/50
171	0.58	75	0.71	--	27/60
171A	0.58	--	0.708	126	27/60
181	0.58	55	0.[illegible]	1[illegible]	27/[illegible]
150A	0.59	--	0.[illegible]	--	--
			0.197	92	
99	0.59	-7	(0.197)	(117)	Unaged
39	0.59	5[illegible]	0.60	1[illegible]	75
98	0.60		0.4[illegible]	1[illegible]	Unaged
1[illegible]	0.60	8[illegible]	0.14	97	[illegible]
154	0.61	74	0.51	94	0
137	0.61	58	0.62	109	
11[illegible]	0.61	64	0.64	135	0
157	0.61	66	0.60	141	15/15[d]
1[illegible]	0.61	70	0.76	13[illegible]	27/7
153	0.61	69	0.59		27/50
155	0.61	69	0.64	130	27/[illegible]
185	0.61		Flaw	Flaw	27/[illegible]
159	0.61	71	0.71	126	27/40
131	0.61	75	0.76	133	27/4[illegible]
114	0.61	73	0.62	96	27/50
182	0.61	76	0.59	12[illegible]	27/5[illegible]
134	0.62	77	0.65	131	27/50
107	0.61	77	0.72	131	27/50
135	0.61	79	0.67	129	27/60
156	0.61	77			27/60
171	0.61	81	0.56	123	27/60
158	0.61	87	0.60	128	27/70
35	0.62	75	0.45	113	0
3	0.63	--	0.58	129	0
2	0.64	--	0.59	13[illegible]	Unaged
1[illegible]	0.64	--	0.58	116	60
10[illegible]	0.64	73	0.55	119	7[illegible]
10	0.65	--	0.63	133	[illegible]
115	0.65	57	0.51	13[illegible]	75

a 0 denotes aged without load.
d First number denotes stress during first 1/2 hour when temperature rises, second number the stress at aging temperature.

TABLE XVd. BARS AXIALLY COMPRESSED AT 75 F.
TENSION TESTS AT 75 F.
E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension During Aging ksi
113	0.67	67	0.21 (0.73)	109 (112)	0
13	0.67	--	0.56	136	0
128	0.67	--	0.79	136	27/30
175	0.67	69	0.37	110	27/41
111	0.67	--	0.40	120	27/50
106	0.67	78	0.01	90	60
105	0.67	32	0.01	94	60
116	0.67	--	0.01	93	27/60
8	0.68	--	0.15	166	0
23	0.70		Fracture at threads		
28	0.70		Fracture at threads		

TABLE XVIa. BARS AXIALLY COMPRESSED AT 75 F.
TENSION TESTS AT -16 F.
E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension During Aging ksi
42	0	46	-	-	0
43	0	51	0.93	125	0
44	0	36	0.93	126	Unaged
45	0	35	0.93	120	Unaged
262	0.20	43	1.05	148	Unaged
263	0.20	37	1.07	149	Unaged
264	0.20	57	0.86	158	0^a
265	0.70	56	0.95	150	0
266	0.70	58	0.89	145	$27/30^d$
267	0.70	60	0.97	153	27/30

a 0 denotes aged without load
d First number denotes stress during first 1/2 hour when temperature rises, second number the stress at aging temperature.

TABLE XVIb. BARS AXIALLY COMPRESSED AT 75 F.
TENSION TESTS AT -16 F.
E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension During Aging ksi
268	0.20	61	0.85	137	27/30
269	0.20	55	0.66	[illegible]	27/35
270	0.20	68	0.92	141	27/40
271	0.20	62	0.87	154	27/60
272	0.30	[illegible]	0.9[illegible]	[illegible]	Unaged
273	0.30	57	0.87	[illegible]	Unaged
274	0.30	61	0.85	136	Unaged
275	0.30	60	0.97	[illegible]	0
276	0.30	62	0.90	150	0
277	0.30	61	0.92	[illegible]	27/30
278	0.30	64	0.90	142	27/30
279	0.30	67	0.89	136	27/30
280	0.30	70	0.91	[illegible]	27/60
281	0.30	[illegible]	0.90	142	27/60
282	0.40	58	0.90	[illegible]	Unaged
283	0.40	53	0.91	116	Unaged
284	0.40	63	0.96	130	0
285	0.40	63	0.86	135	0
286	0.40	[illegible]	0.83	141	27/30
287	0.40	66	0.85	152	27/30
288	0.40	69	0.91	140	27/50
289	0.40	70	0.81	[illegible]	27/50
290	0.40	60	0.94	147	27/60
291	0.40	[illegible]	0.90	142	27/60
225	0.50	[illegible]	0.87	[illegible]	Unaged
227	0.50	47	0.86	144	Unaged
210	0.50	67	0.82	159	0
217	0.50	65	0.83	146	0
263	0.50	62	0.89	148	27/15
223	0.50	68	0.85	163	27/30
228	[illegible]	68	0.87	[illegible]	27/30
226	0.50	69	0.57	130	27/30
214	0.50		0.67		27/40
218	0.50	67	0.86	[illegible]	27/50
[illegible]	0.50	[illegible]	0.83	[illegible]	27/50
[illegible]	0.50	[illegible]	[illegible]	[illegible]	27/50
[illegible]	0.50	[illegible]	[illegible]	[illegible]	27/60
[illegible]	0.50	[illegible]	[illegible]	[illegible]	27/60
213	0.50	67	0.67	[illegible]	27/70
[illegible]	0.50	[illegible]	0.82	[illegible]	27/70
250	0.55	[illegible]	0.87	155	Unaged

TABLE XVIc. BARS AXIALLY COMPRESSED AT 75 F.
TENSION TESTS AT -16 F.
E-STEEL.

Bar E-Steel	Prestrain	0.1% Offset Stress ksi	Fracture Nat. Strain	Fracture True Stress ksi	Tension During Aging ksi
[illegible]	[illegible]		[illegible]	[illegible]	$Unaged^a$
[illegible]	[illegible]		[illegible]	[illegible]	
[illegible]	0.58	Flaw	[illegible]	Flaw	0
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	0
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]^d
251	[illegible]	69	1.01	[illegible]	[illegible]
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/30
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/30
191	0.58	73	[illegible]	[illegible]	27/40
[illegible]	[illegible]		[illegible]		27/40
[illegible]	[illegible]	-	[illegible]	[illegible]	[illegible]
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
[illegible]	0.58	75	0.70	[illegible]	27/50
[illegible]	[illegible]	78	0.72	143	27/60
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/60
[illegible]	0.58	[illegible]	[illegible]	[illegible]	27/70
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
156	0.61	60	1.04	[illegible]	Unaged
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	Unaged
[illegible]	0.61		[illegible]	[illegible]	Unaged
[illegible]	0.61		[illegible]	[illegible]	Unaged
[illegible]	[illegible]	67	[illegible]	[illegible]	0
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	0
[illegible]	[illegible]		[illegible]	[illegible]	0
[illegible]	[illegible]		[illegible]	137	0
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/13
[illegible]	[illegible]		[illegible]		27/27
[illegible]	0.61	73	[illegible]	[illegible]	27/30
[illegible]	0.61	75	[illegible]	142	27/30
135	[illegible]	72	[illegible]	136	27/40
[illegible]	0.61		[illegible]	92	27/40
[illegible]	0.61		[illegible]		[illegible]
[illegible]	0.61	[illegible]	[illegible]	105	27/40
254	0.61	73	[illegible]	[illegible]	27/50
198	[illegible]	[illegible]	[illegible]	137	27/50
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/60
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	27/60
[illegible]	0.61	[illegible]	[illegible]	[illegible]	27/70
[illegible]	0.61	[illegible]	[illegible]	[illegible]	27/70

a 0 denotes aged without load
d First number denotes stress during first 1/2 hour when temperature rises, second number the stress at aging temperature.

having as-rolled surfaces, which is not inconsistent with the increase of the exhaustion limit by about 0.05 found for bars with machined surfaces (67). The remaining difference may be attributed to the post-compression machining of the axially compressed bars instead of the pre-compression machining of the reversed-bend bars. Post-compression machining should eliminate any small surface irregularities developing during compression, and should increase the apparent overall ductility.

The 0.1% offset yield strength, the true stress, and the natural strain at fracture have also been plotted according to aging treatment, for compressive prestrains of 0.48-0.52, 0.54-0.58, 0.59-0.63, and 0.65-0.67 for tests at 75 F (Figures 33a and 33b); and likewise for prestrains of 0.50, 0.58, and 0.61 at -16°F (Figure 34). Contrary to earlier expectation it is not possible to distinguish any decisive trend, except a small increase of the 0.1% offset strength with the aging tension. The 0.1% offset strength is always appreciably higher than the aging tension. In unstrained bars tested at -16°F, however, aging has a considerable effect on the lower yield point. As shown by the first four lines of Table XVI, aging without load raised the lower yield point from 36 ksi to 46 and 51 ksi, but did not have a significant influence on the natural fracture strain or on the true fracture stress. All four bars showed an upper yield point. For better comparison aged bar A2 and unaged bar U4 were cut from a single 9" long bar, as also were bars A3 and U5.

8. CONCLUSIONS

A consideration of the relation between local strains at a crack and the behavior of a structure as a whole has led to a definition and a criterion of brittleness of failure for a structure based on the average net stress at fracture. It then became evident that mild structural steel has sufficient ductility to avoid brittle fracture under static loading even when it has the deepest notches and a low temperature, and that low stress fractures under static load should result

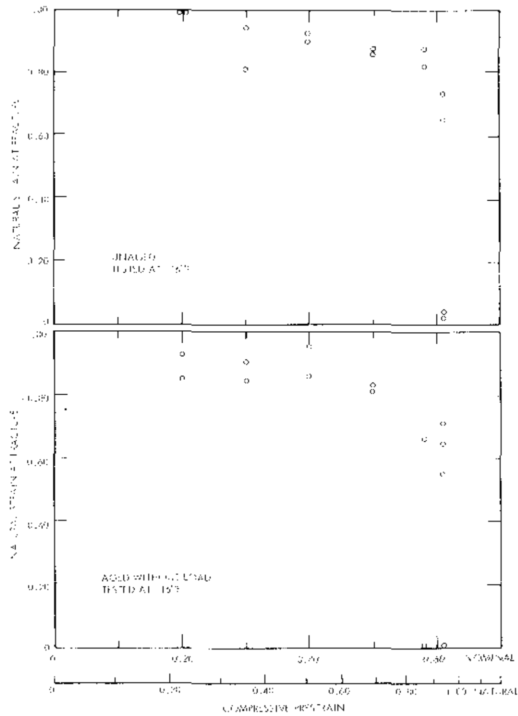


FIG. 31. NATURAL STRAIN AT FRACTURE VS. PRESTRAIN OF BARS OF E-STEEL, UNAGED AND AGED WITHOUT LOAD AND TESTED AT -16 F.

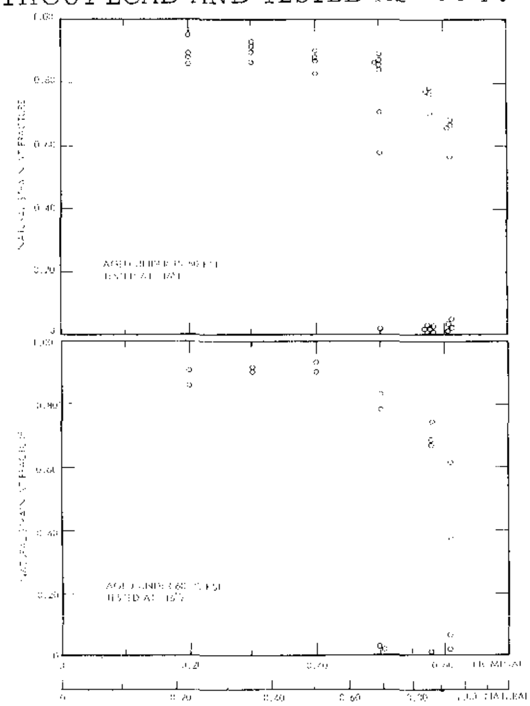
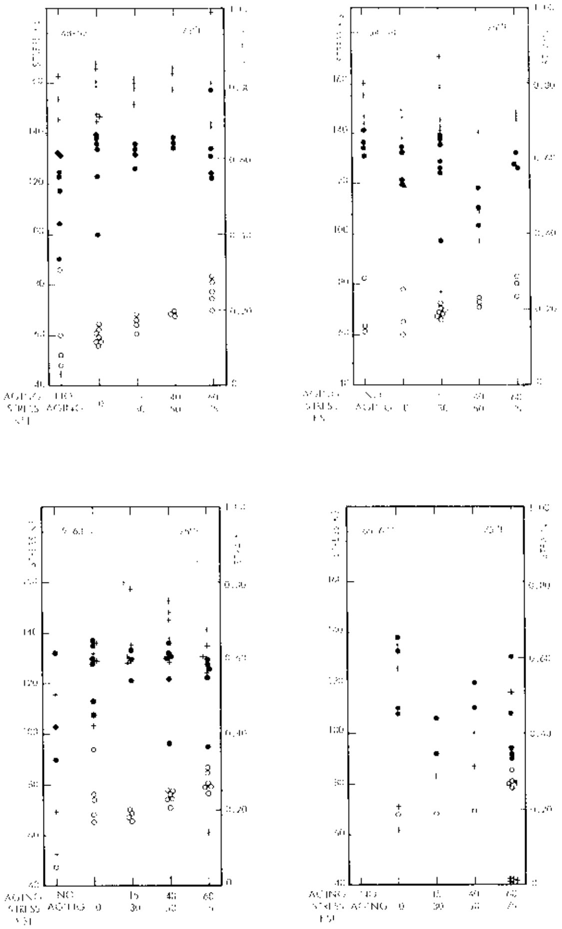


FIG. 32. NATURAL STRAIN AT FRACTURE VS. PRESTRAIN OF BARS OF E-STEEL, AGED UNDER 15-50 AND 60-75 KSI AND TESTED AT -16 F.



(a)

(b)

FIG. 33. 0.1% OFFSET YIELD STRENGTH (o); TRUE STRESS (●); AND NATURAL STRAIN (+) AT FRACTURE OF PRECOMPRESSED BARS OF E-STEEL TESTED AT 75 F.

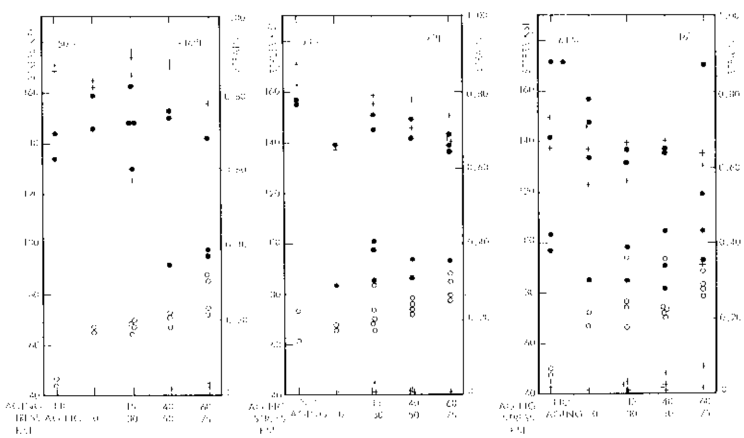


FIG. 34. 0.1% OFFSET YIELD STRENGTH (o); TRUE STRESS (●); AND NATURAL STRAIN (+) AT FRACTURE OF PRECOMPRESSED BARS OF E-STEEL TESTED AT -16 F.

from a prior suitable reduction or exhaustion of the initial ductility of structural steel.

The tests which were then undertaken have demonstrated the important influence of the whole history of strain and temperature on the subsequent ductile or brittle behavior of steel. The magnitude of some of these effects had not been widely recognized. The extensive tests with notched plates, bent bars, and axially compressed bars of E- and other steels, subjected to various types of straining and heating history, have shown the following:

A. Sufficient precompression at room temperature exhausts the extensional ductility to such an extent as to result in fracture after extensions of the order of 0.01 (1%).

B. The reduction of extensional ductility is not proportional to the amount of precompression. A narrowly determined limit of compressive prestrain exists (the exhaustion limit) at which the ductility suffers a rapid reduction. Prestrains smaller than this limit have little effect on the extensional ductility. Larger prestrains cause complete embrittlement. The exhaustion limit was found to be about 0.5 (50%).

C. The exhaustion limit is lower (easier embrittlement) when the final test is made at a lower temperature.

D. The exhaustion limit is lower for bars which are subsequently aged (i.e. heated for 90 min. at about 300°F).

E. Exhaustion of ductility by precompression and aging does cause brittle fracture in originally ductile plates. Notched plates sufficiently precompressed in their plane at right angles to the notch axis suffered such a reduction of ductility in the highly plastically compressed notch regions, so as to fracture or develop arrested cracks under static loading, at average net stress levels as low as 10% of virgin yield. These tests duplicate service fractures insofar as stress level, fracture appearance and absence of plastic deformation are concerned.

F. Exhaustion of ductility is highly anisotropic, as was conclusively demonstrated by Allen (31) and Rendall (32), and had been indicated by the transversely prestrained notched plate tests (2). Steel compressed to the extent to be brittle in tension in the same direction as the compression, was highly ductile in a transverse direction. Likewise cold rolled steel has much less ductility in the direction of the thickness than in the rolling direction.

G. Localized residual stresses of yield intensity have no influence on the brittle or ductile behavior of the test specimens.

H. An increase of prestraining temperature reduces the exhaustion limit (easier embrittlement) for subsequent tests at 75° or -16°F. The reduction is highest at about 500-600°F, where the exhaustion limit has about half the value for cold prestraining. Prestraining above 700° raises again the exhaustion limit, till at about 900° it is higher than at room temperature (more difficult embrittlement). Hot straining in extension can also exhaust the ductility in subsequent cold tension.

I. Heat treatment at temperatures about 1100°F, of a duration dependent on the amount of prestrain, can restore sufficient ductility to prevent fracture at small strains. The effectiveness of the so-called "thermal stress relieving" operation in preventing brittle fracture appears to result mainly from the restoration of ductility than from the reduction of the residual stresses.

Exhaustion of ductility by hot or cold straining and aging appears as a fundamental cause of brittle fracture initiation. This is in agreement with the observation (65) that the origin of service fractures is usually traced to a cold-worked region or to a defect close to a weld, where complex hot straining occurs during the welding cycle. It is quite likely that very local strains of the order of 0.50 may develop at a strain concentration or at a defect in such a region, and more so in regions which have been work hardened by external means. It appears even likelier that local strains of the order of 0.20 develop at a defect close to a weld at temperatures around 500°F. In effect the weld zone is known to suffer a uniform stretching of the order of 0.02 during cooling, and the concentration due to a defect could raise this substantially. The straining of this region is quite complicated because longitudinal and transverse compression develops as the weld approaches, and is changed into tension when it recedes. However, there is no indication that the cold or hot straining now employed are the most effective methods of embrittlement or the closest to reality. A strain and temperature history causing easier embrittlement may exist.

It may also be found that prestrained steel which is still ductile in simple tension is brittle under the triaxial constraint of a notch. This shows how much has still to be learned about the strains developing at cracks or notches in materials of specific anisotropic strain hardening laws, and on the behavior of prestrained steel under the specific conditions of strain and constraint existing at a notch.

The present investigation, however, has clearly shown that the properties responsible for brittle fracture are those of damaged and not of virgin steel. It would appear then that tests designed to measure the properties of embrittled steel, or better, the ease with which steel is embrittled during some straining cycle, would give results more directly related to brittle fracture than tests of virgin steel. It is therefore suggested that tests of embrittlement by cold or hot straining be seriously considered as possible acceptance tests for steel. The reversed bend test in particular provides a practical reproducible test for the study of the amount of strain which causes embrittlement, and of the effect of aging, of the temperature of straining or testing, of surface condition, etc. It offers real hope of becoming also a simple shop and field inspection test of direct and understandable physical meaning.

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