SSC-185

EFFECT OF SURFACE CONDITION ON THE EXHAUSTION OF DUCTILITY BY COLD OR HOT STRAINING

by

J. Dvorak and C. Mylonas

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ADDRESS CORRESPONDENCE TO:

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July 1968

Dear Sir:

Determining the effect of gross strain upon the mechanical and metallurgical properties of steel has been a subject of study by the Ship Structure Committee at Brown University in a project called "Macrofracture Fundamentals". Herewith is a copy of *Effect of Surface Condition on the Exhaustion of Ductility by Cold or Hot Straining* by J. Dvorak and C. Mylonas, which describes a portion of the study.

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

This report is being distributed to individuals and groups associated with or interested in the work of the Ship Structure Committee. Comments concerning this report are solicited.

Sincerely yours,

D. B. Henderson

D. B. Henderson
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

SSC - 185

Seventh Progress Report

on

Project SR - 158

"Macrofracture Fundamentals"

to the

Ship Structure Committee

EFFECT OF SURFACE CONDITION ON THE EXHAUSTION OF DUCTILITY BY COLD OR HOT STRAINING

by

J. Dvorak and C. Mylonas

Brown University
Providence, R. I.

under

Department of the Navy
Naval Ship Engineering Center
Contract Nobs 88294

U. S. Coast Guard Headquarters
Washington, D. C.

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ABSTRACT

The compressive prestrain (exhaustion limit) needed to cause brittle behavior in subsequent tension was found to be much higher in ABS-B steel bars with surfaces machined by about 0.030 in. before straining than with as-rolled surfaces, even more so when the surfaces were machined after straining. Removal of the strained surface caused a small increase of exhaustion limit even when the surfaces had been machined before prestraining. In all cases the increase was larger for bars prestrained at 550°F than at 70°F. The surface effect was found stronger than in earlier tests with an ABS-C steel. In addition the microhardness was found to rise gradually in a 0.030 in. layer adjacent to the surface and to reach a peak at the surface itself in all as-rolled or as-strained surfaces.

The surface damage from an unfavorable rolling history permits an easier surface embrittlement by hot straining in a region of strain concentration close to a weld and creates a dangerous trigger of brittle fracture, as is indicated by service fractures starting at such regions. A study of the rolling and straining history causing such weak regions could help their prevention.

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1. INTRODUCTION

The importance of the history of strain and temperature and of the final state of stress on the subsequent ductility of mild steel has been described in several earlier papers, summarized and extended in references [1-5], which contain many related references. The change of ductility in simple tension after precompression in the same direction is of special interest. At first the ductility (i.e. the strain at fracture) is high and remains essentially unchanged up to a compressive prestrain of the order of 0.50 (i.e. 50%). At higher prestrains the ductility is rapdily reduced to levels as low as 0.01 within a relatively narrow range of prestrains, which has been called the exhaustion limit for the particular conditions of prestraining and testing. A much narrower transition range, hence a better defined exhaustion limit has been found with bars compressed and extended by a reversed bending action than in axial compression-tension [4-6].

The exhaustion limit is a measure of the resistance of the steel to embrittlement under the specific conditions of prestraining in compression followed by a reversal to simple tension. As should be expected the tougher steels were generally found to have a higher exhaustion limit [6], i.e. to resist embrittlement more than less tough steels. The contributory action of other embrittling factors, such as accelerated aging after prestraining and low testing temperatures, caused a reduction of the exhaustion limit, i.e. made embrittlement by prestraining easier.

The temperature of prestraining was found to have a strong influence on the ductility in tension at 70^{0} or -16^{0}F [7,8]. As the straining temperature increased up to about 600^{0}F, the exhaustion limit gradually decreased to about one half the limit for straining at 70^{0}F. Straining at still higher temperatures gradually raised again the exhaustion limit. Suitable heat treatment at 700^{0} to 1200^{0}F after embrittlement by prestraining was found to restore ductility [8]. As was expected the required heat treatment was shortened when the temperature was increased. It was also found, however, that the duration of heat treatment at a fixed temperature increased rapidly with the amount of prestrain: it became impractically long beyond a limiting prestrain unless a higher temperature were used. Restoration of ductility was always considerably easier (faster or at lower temperature) after embrittlement by hot than by cold straining.

Another aspect of the embrittlement caused by precompression is its pronounced anisotropy, as was clearly demonstrated by Allen [9]. Highly compressed iron and steel can be brittle in tension in the same direction as the precompression, but highly ductile in a transverse tension. Likewise a large reduction by cold rolling may cause brittleness in tension in the direction of reduced thickness but not in the direction of rolling.

The reduction of ductility in steel subjected to cold or hot precompression was found by Körber, Eichinger and Möller in the early 1940's [10], but does not appear to have been directly connected with the problem of brittle fracture in service. It appears to have passed completely unnoticed and was rediscovered 15 years later in connection with brittle fracture initiation in steel [2-6] which it qualitatively explains. In effect fracture initiation in service has been usually traced to regions of stress concentration which, in addition, had been cold worked or had been deformed hot, as e.g. at defects or re-entrant corners close to welds. The influence of the history of strain and temperature on the ductility under the local conditions of stress at a crack or notch appears as a fundamental factor in brittle fracture initiation in mild steel under static loading. The other factor is the required straining under the local stress tensor up to large overall loads, as is discussed in earlier papers (1-4, 11).

2. PURPOSE OF THE TESTS

Considerable interest has been focussed on the reduction of the initial ductility of mild steel by suitable straining, because this funda-

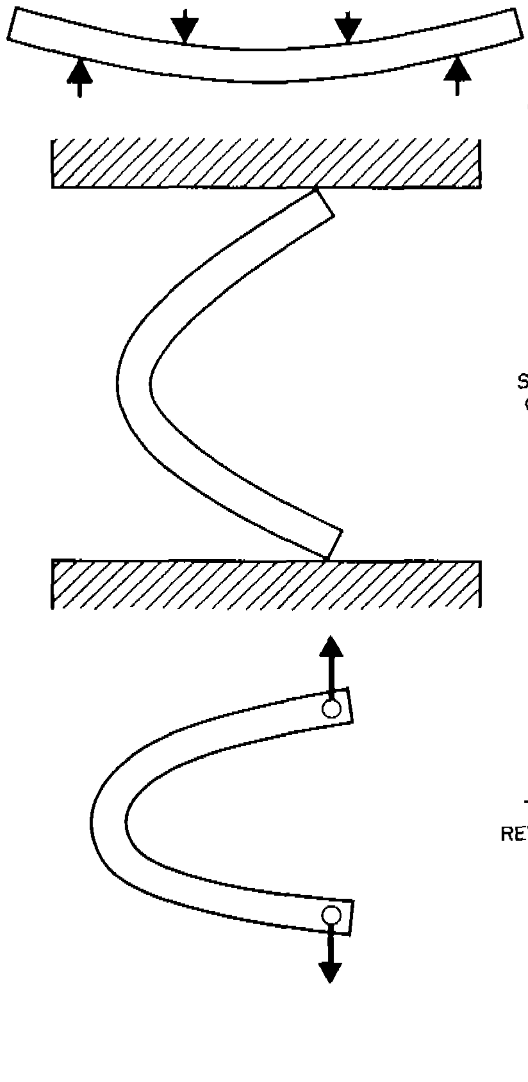


FIG. 1a
FIRST STAGE
OF BENDING

FIG. 1b
SECOND STAGE
OF BENDING

FIG. 1c
THIRD STAGE
REVERSED BENDING

mental cause of brittle fracture can be reproduced consistently under controlled laboratory conditions. It may be said that a suitably prestrained piece holds the key to the understanding of brittle fracture initiation. An explanation of the mechanism which causes suitably strained steel to behave in a brittle manner under the local condition of a crack or notch may lead to its prevention or to the selection of less easily embrittled steels. The influence of the various modifying factors such as type and temperature of prestrain or testing, aging, heat treatment, or of local stress tensor are very important. At the very first, however, it is indispensable to know as accurately as possible the conditions causing embrittlement. The present tests were made in order to investigate an apparent discrepancy between exhaustion limits found by different methods. The nominal prestrain at the exhaustion limit of ABS-B steel compressed axially at $70^{\circ}F$ and subsequently tested in simple tension at $-16^{\circ}F$ was found to be about 0.75 (11). The same steel precompressed (on the one side) by bending and tested in reversed bending (Fig. 1) gave an exhaustion limit of only 0.48 (8). The corresponding natural strains are -1.39 vs. -0.65 giving the striking difference of 0.74.

Of course the two sets of results are not directly comparable because the surfaces where brittle fracture starts are different. The reversed bending tests were made with bars having the initial as-rolled surface of the steel plate, whereas in the axial tests the bars after compression were machined to standard 0.505 in. diameter tension specimens. In an earlier study reversed bend tests were made with bars with a machined surface [12]. The exhaustion limit of an ABS-C steel prestrained at $70^{\circ}F$

and tested at -16^{o}F was raised from about 0.56 with as-rolled surfaces to 0.60 with machines surface, i.e. an increase of only 0.04. Axial tests with ABS-C steel were not made, but its exhaustion limit should be at least as high as for ABS-B steel, i.e. about 0.75. The major difference between exhaustion limits by reversed bending and axial tests remained unexplained.

The surface condition during final testing is different in the reversed bend and axial tests even when bars with initially machined surfaces are used. In the final phase (tension) the specimens tested in reversed bending have as-strained surfaces; in axial testing they have newly machined surfaces. It was decided to check whether the influence of the as-strained surface would explain the difference between reversed bend and axial tests. It was considered especially interesting to examine the possible existence of such a surface effect also in hot prestraining, as it might be related with the considerably lower exhaustion limit of bars prestrained at 550^{o}F and tested in reversed bending at -16^{o}F as compared with bars prestrained at 70^{o}F, namely about 0.23 vs. 0.48 respectively with ABS-B steel [8].

This study required comparative tests between bars prestrained and tested with as-strained surfaces and with surfaces machined after straining. Obviously axial tension tests could only be made with bars having machined surfaces because their test section has to be made smaller than the pulling heads. Accordingly most tests were made in reversed bending and were compared with axial tests. For the sake of brevity the symbol AR indicates as-rolled, M machined, AS as-strained surfaces, and numbers indicate the temperature of prestraining in ^{o}F. All final tests in reversed straining were made at $-16^{\circ}F$. The following tests were made in reversed bending unless otherwise specified:

a. AR(70)AS: Bars with as-rolled surfaces prestrained at $70^{\circ}F$ and tested in the as-strained state.

b. M(70)AS: Machined, prestrained at $70^{\circ}F$, tested as-strained.

c. M(70)M: Machined, prestrained at $70^{\circ}F$, re-machined and tested.

d. AR(550)AS: Prestrained as-rolled at $550^{\circ}F$, tested as-strained.

e. AR(550)M: Prestrained as-rolled at $550^{\circ}F$, machined and tested.

f. M(550)AS: Machined, prestrained at $550^{\circ}F$, tested as-strained.

g. M(550)M: Machined, prestrained at $550^{\circ}F$, re-machined and tested.

h. M(550)M-Axial: Axially compressed at $550^{\circ}F$, machined, tested in tension.

i. Microhardness tests at various depths of bars prestrained at 70° and $550^{\circ}F$.

j. Microetch study of various sections of bars prestrained at $70^{\circ}F$ and $550^{\circ}F$.

3. REVERSED-BEND TEST PROCEDURE

ABS-Class B steel plate $\frac{3}{4}$ in. thick was used throughout. It was part of the same heat used in previous tests at the National Bureau of Standards, from which the typical compositions and properties shown in Tables I and II are taken. Plate 7IN was used in all preliminary reversed bend tests and in axial tests. Plate 77N was used in the main reversed bend tests. The length of all bars coincided with the direction of rolling, except in a few bars used to check the possible effect of a transverse rolling direction. The dimension of the bars used in reverse-bend tests were about 0.750 x 1.000 x 8.125 in., but for the highest prestrains the length was reduced to 7.375 in. Bending of the bars was in a plane perpendicular to the original plate surface. No heat treatment was used prior to prestrain-

TABLE I COMPOSITION OF ABS-B STEEL.

	C	Mn	P	S	Si	Ni	Cu	Cr	Al	N
Minimum	0.14	0.91	0.009	0.018	0.041	0.021	0.051	0.023	0.02	0.004
Maximum	0.18	1.07	0.012	0.028	0.056	0.040	0.096	0.031	0.03	0.006
Typical	0.14	1.04	0.011	0.018	0.056	0.023	0.083	0.031	0.02	0.004
	0.15	0.94	0.009	0.027	0.046	0.040	0.094	0.023	0.02	0.005

TABLE II PROPERTIES OF ABS-B STEEL.

	Yield Point ksi	Ultim. Strength ksi	Elong. (8") %	Finish Temp. °F	Ferrite Grain Size	°F			Nil Duct. Temp. °F Center	Fibrous	
						T_{V10}	T_{V15}	T_{V20}		50% °F	10% °F
Maximum	32.6	57.9	31.0	1600	7.8	-30	-24	-13	-20	24	-22
Minimum	35.7	63.9	33.0	1725	8.2	-5	6	18	-10	39	-10
Typical	33.8	58.4	33.0	1640	7.8	-5	6	18	-10	37	-14
	35.7	59.8	32.0	1600	8.1	-11	2	+11	-10	28	-15

From 12 analyses and 6 tests by the Nat. Bureau of Standards on pieces taken from plates of the same heat as used in the present tests.

ing at 70°F or at 550°F (the temperature of lowest exhaustion limit [8]) henceforth referred to as cold and hot straining. The cold strained bars were subjected to an accelerated aging of 2 hours at 300°F. Bars prestrained hot in bending were cooled in air; axially prestrained bars were cooled in boiling water because axial compression was much slower than bending, and it was desirable to reduce the total time of each bar at elevated temperature. All fracture tests were made at -16°F. The reversed-bend tests were carried out in three stages (Fig. 1). As already mentioned and described in detail in references [5-8] the ductility in tension during the reversal of bending remains quite high up to the narrow prestrain range of the exhaustion limit, at which it suddenly droped to values of the order of 0.02. The interesting characteristic of this test is that the drop of ductility is reflected in a drop of the load at fracture. The prestrain range at which the fracture load is rapidly reduced gives the exhaustion limit (e.g. Fig. 7) without the need of any strain measurement at fracture.

Removal of a surface layer of the bent bars was more difficult than with straight bars. The intrados of the bent bars was saddle-shaped (Fig. 4) with an anticlastic surface resembling part of the interior of a toroid. Machining of the surface was done in a matching toroid shape by side-milling with a cutter of suitable diameter while the specimen was held in an indexing head (Fig. 2), and was swung about an axis in the

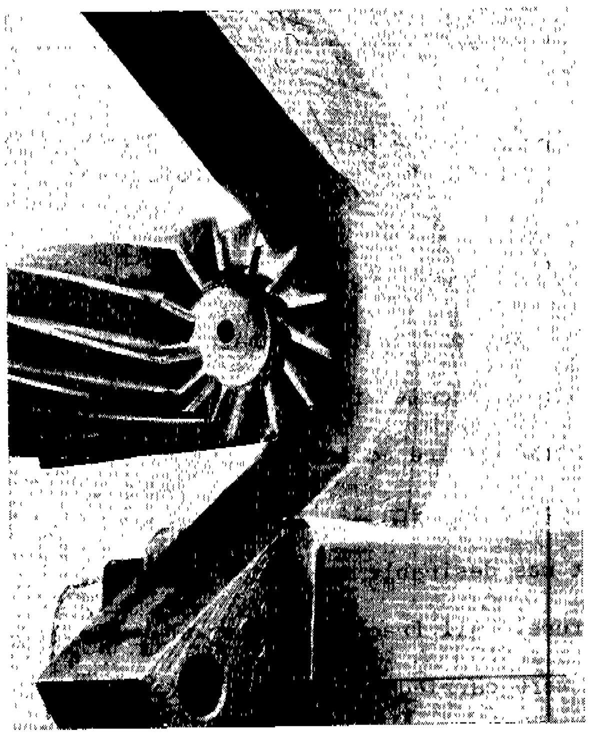


Fig. 2 Machining of Bent Bars.

mid-plane of the bent bar through the center of transverse curvature (in Fig. 2 the axis is vertical just beyond the extrados). Machining was of the required depth at the cross-section of highest curvature where fracture normally occurs and tapered off in the longitudinal direction on both sides, as shown in Figs. 2 and 3. In the transverse direction the machining depth increased from the center line to the lateral surfaces (Fig. 4), so that the middle of the curved surface has always the highest prestrain. The removed thickness Δh at the center of the most curved part is always reported in the test results. Several measurements of Δh were taken and averaged because the scatter occasionally reached as much as ± 0.004 in.

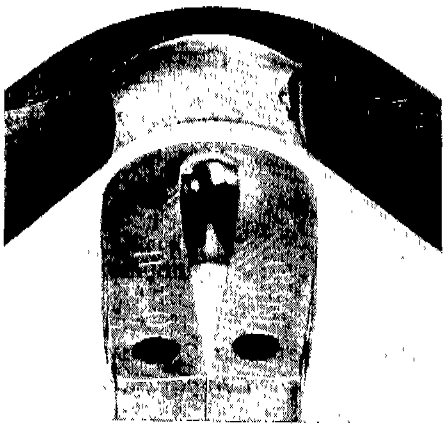


Fig. 3 Bars Machined After Prestrain Top: After 0.30; Bottom: After 0.71.

The nominal prestrain ε of the intrados after bending was found (5,6) from the bar thickness h and the minimum radius of curvature R

$$\varepsilon = h/(2R+h)$$

The maximum prestrain of bars machined after bending was found from the strain ε after an approximate correction $\Delta\varepsilon$

$$\Delta\varepsilon = 2\varepsilon\Delta h/h,$$

based on an assumption of plane cross-sections, of a neutral plane at mid-thickness and of negligible influence of curvature and large strains. All these assumptions together may at most introduce a small error in the already small correction $\Delta\varepsilon$. The prestrain reported in all results with bars machined after straining is the quantity $(\varepsilon-\Delta\varepsilon)$.

5. RESULTS OF REVERSED BEND TESTS

The main series of tests were preceded by a preliminary series which indicated the required range of prestrains and confirmed that the removal of the

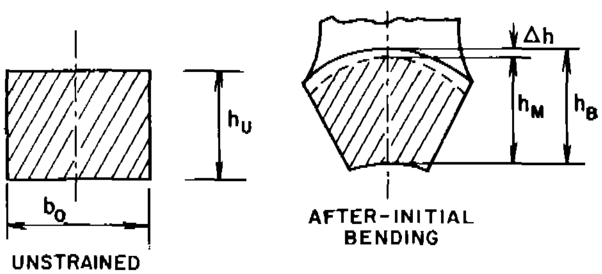


Fig. 4 Cross Sections Of Reversed - Bend Bars (Dashed Line Indicates Machined Surface).

surface caused a considerable increase in the exhaustion limit, far more than expected on the basis of earlier tests with ABS-C steel (12), especially for hot-strained bars. The preliminary results are included in the main series.

The exhaustion limits of bars with as-rolled surfaces perstrained at $70^{\circ}F$ or $550^{\circ}F$ and tested at $-16^{\circ}F$ were first determined (Tables III, IV and

TABLE III AR(70)AS ABS-B STEEL REVERSED-BEND TESTS. BARS WITH AS-ROLLED SURFACES PRESTRAINED AT 70°F, AGED AND TESTED AT -16°F.

BAR	PRESTRAIN	FRACTURE LOAD (lb.)		FRACTURE STRESS $4M/b_oh^2$ (ksi)		BAR SIZE
		1st CRACK	FRACT.	1st CRACK	FRACT.	
B-697	0.30	-	> 7500	-	> 90	0.785 x 1.000 x 7.375 in.
B-698	0.31	-	> 7500	-	> 91	
B-695	0.33	-	> 7500	-	> 92	
B-696	0.33	-	> 7500	-	> 82	
B-691	0.36	-	> 7500	-	> 93	
B-692	0.36	-	> 7500	-	> 95	
B-694	0.41	-	> 7500	-	> 98	
B-693	0.43	-	3700	-	59	
B-622	0.44	-	> 7500	-	>105	
B-625	0.44	-	> 7500	-	> 98	
B-624	0.44	100	2000	2	-	
B-626	0.44	100	2000	2	-	
B-607	0.45	-	2000	-	36	
B-608	0.45	-	> 7000	-	>130	
B-623	0.46	-	2500	-	40	
B-609	0.49	-	4200	-	72	
B-610	0.49	-	2100	-	36	
B-611	0.51	100	2000	2	-	
B-612	0.51	1000	2800	18	-	

Figs. 5,6). The exhaustion limits were then found for bars with surfaces machined by 0.035 or 0.040 in. and then subjected in the as-strained condition to the same reversed-bend test (Tables V, VI and Fig.s 5, 7). The influence of the surface layer was studied with as-rolled hot strained bars subsequently machined on the surface, as shown in Figs. 2 and 4, to three depths of approximately 0.014, 0.026 and 0.046 in. (Table VII and Fig. 8). Finally tests were

TABLE IV AR(550)AS ABS-B STEEL REVERSED-BEND TESTS. BARS WITH AS-ROLLED SURFACES PRESTRAINED AT 550°F, TESTED AT -16°F.

BAR	PRESTRAIN	FRACTURE LOAD (lb.)		FRACTURE STRESS $4M/b_oh^2$ (ksi)		BAR SIZE
		1st CRACK	FRACT.	1st CRACK	FRACT.	
B-721	0.21	-	> 8500	-	80	0.785x1.000x7.375 in.
B-722	0.21	-	> 8500	-	80	
B-639	0.22	-	> 7500	-	78	
B-712	0.22	-	> 8500	-	89	
B-640	0.22	-	6000	-	60	
B-711	0.22	500	2000	6	-	
B-641	0.27	-	> 7500	-	85	
B-642	0.27	3000	4000	39	-	
B-709	0.34	500	2000	7	-	
B-710	0.34	500	2000	7	-	

made with bars machined both before and after prestraining in order to study the effect of straining on the free surface (Table VIII and IX and Fig. 9). The results are compared in Table XI and Fig. 10 (which also contains a curve of exhaustion limit vs. prestrain temperature [8]) and may be summarized as follows:

a. Bars prestrained at 550°F or at 70°F and tested at -16°F have the lowest exhaustion limit when tested with as-rolled surfaces.

b. Machining of the surface up to a depth of about 0.040 in. before straining raises considerably the exhaustion limit.

c. Machining by about 0.040 after straining or partly before and at least about 0.010 in. after straining raises the exhaustion limit limit by about 0.25 for bars strained hot and about 0.17 cold.

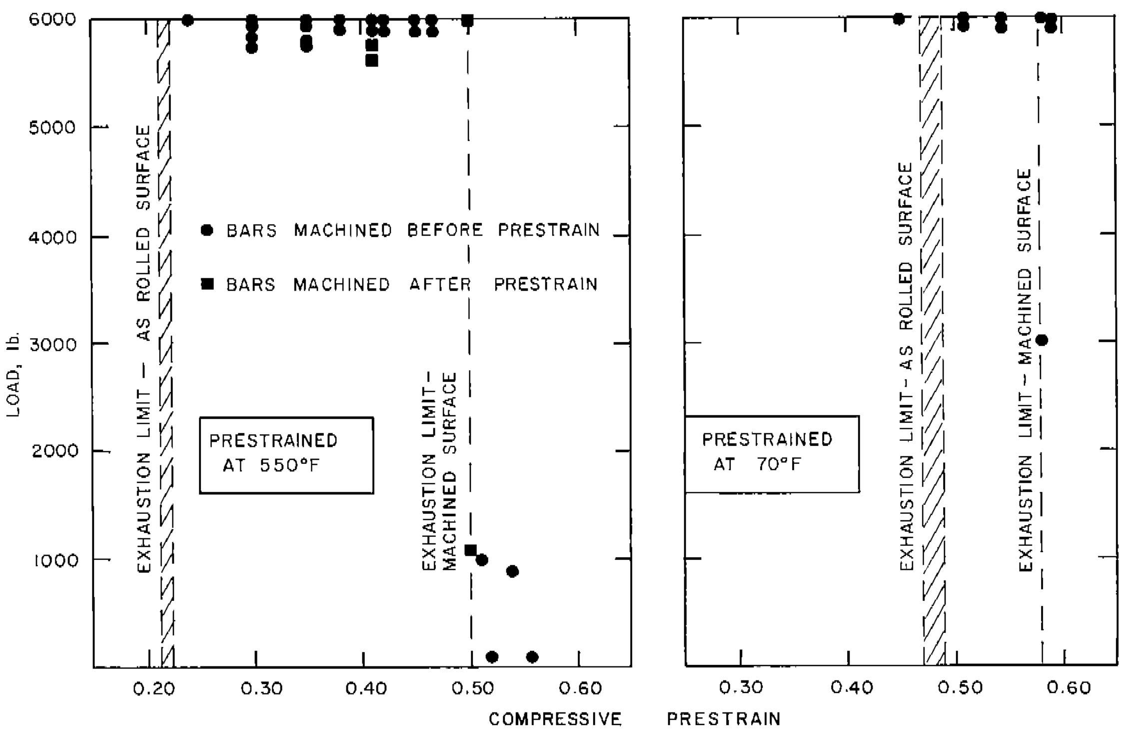


Fig. 5 Preliminary Tests On Effect Of Surface Layer.

d. The exhaustion limits for as-rolled surfaces; machined before; machined after straining are respectively:

for bars prestrained at 550^{O}F: 0.22-0.27; 0.43-0.46; 0.50-0.51

for bars prestrained at 70^{O}F: 0.43-0.45; 0.57-0.59; 0.61

It may be concluded that a surface layer about 0.040 in. deep is responsible for the low resistance to strain embrittlement of as-rolled ABS-B steel bars. Removal of this layer raises considerably the exhaustion limit.

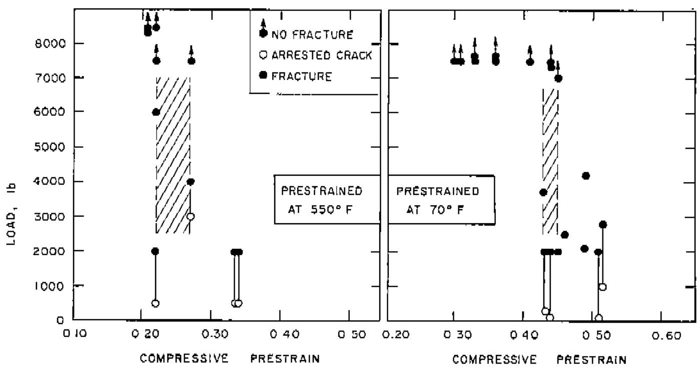


Fig. 6 Reversed-Bend Tests Of Bars With As-Rolled Surfaces.

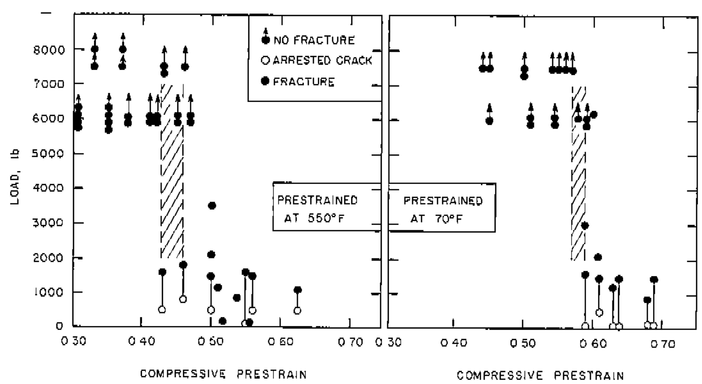


Fig. 7 Reversed-Bend Tests Of Bars Machined Before Prestraining.

Furthermore the small increase of exhaustion limit found when initially machined surfaces were again machined after straining shows that straining damages the free surface more than the interior.

A comparison of the effect of machining the surface before <u>cold straining</u> in ABS-B and an ABS-C steel reported earlier (12) is of interest.

TABLE V M(70)AS ABS-B STEEL REVERSED-BEND TEST. BARS WITH MACHINED SURFACES PRESTRAINED AT 70°F, AGED AND TESTED AT -16°F.

BAR	PRESTRAIN	SURFACE PREPARATION	FRACTURE LOAD (lb.)		FRACTURE STRESS $4M/b_oh^2$ (ksi)		BAR SIZE
			1st CRACK	FRACT.	1st CRACK	FRACT.	
B-682	0.44	BARS MACHINED BEFORE PRESTRAIN 0.040 in. EACH SURFACE	-	> 7500	-	> 122	0.710 x 1.000 x 7.375 in.
B-681	0.45		-	> 7500	-	> 121	
B-683	0.50		-	> 7500	-	> 120*	
B-684	0.50		-	> 7500	-	> 120*	
B-685	0.54		-	> 7500	-	> 116*	
B-686	0.55		-	> 7500	-	> 115*	
B-746	0.56		-	> 7500	-	> 106*	
B-747	0.57		-	> 7500	-	> 110*	
B-688	0.59		50	1600	1	-	
B-687	0.60		-	6200	-	95*	
B-605	0.61		-	2100	-	48	
B-606	0.61		500	1500	12	-	
B-603	0.63		100	1200	2	-	
B-604	0.64		100	1500	2	-	
B-601	0.68		50	900	1	-	
B-602	0.68		100	1500	1	-	
B-552+	0.45		-	> 6000			
B-553+	0.51		-	> 6000			
B-554+	0.51		-	> 6000			
B-557+	0.54		-	> 6000			
B-558+	0.54		-	> 6000			
B-561+	0.58		-	3000			
B-562+	0.58		-	> 6000			
B-563+	0.59		-	> 6000			
B-564+	0.59		-	> 6000			

* Large plastic strains, 0.24 to 0.45, appearance of stars and tiny shear cracks on the surface.

\+ Preliminary results with bars 0.71 x 1.00 x 8.125 in., machined 0.040 in. before prestraining from plate 71N.

The exhaustion limit of unaged ABS-C steel had been found to be 0.56 for as-rolled and about 0.60 for machined surfaces. No results exist for aged ABS-C steel but on the basis of earlier tests it may be estimated that the

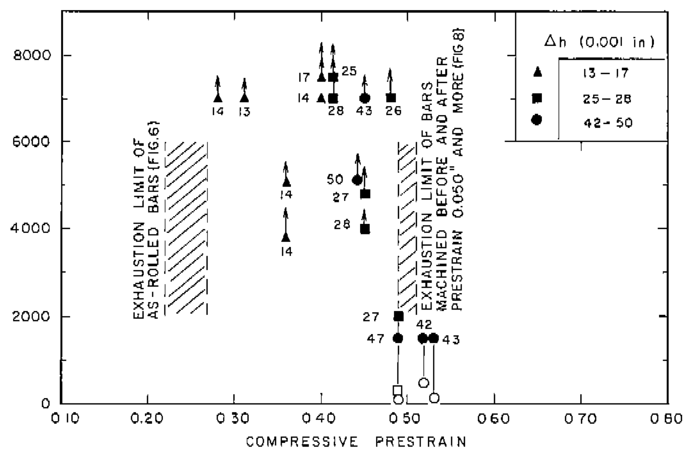


Fig. 8 Reversed-Bend Tests Of Bars Machined By Different Amounts Δh After Prestrain.

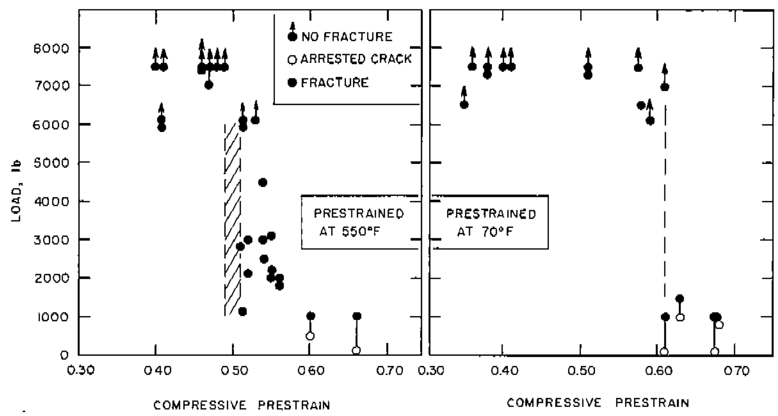


Fig. 9 Reversed Bend Tests Of Bars Machined Before And After Prestraining.

limits would be about 0.06 lower, i.e. about 0.50 for as-rolled and about 0.55 for machined ABS-C steel. The limits for aged ABS-B steel are much further apart: 0.44 for as-rolled and 0.58 for machined surfaces. Thus aged ABS-B and -C steels have similar exhaustion limits for machined surfaces (0.58 vs. 0.55 or a difference of about 0.03) but not for as-rolled surfaces (0.44 vs. 0.50 or an opposite difference of -0.06). The present

TABLE VI M(550)AS ABS-B STEEL REVERSED-BEND TEST. BARS WITH MACHINED SURFACES PRESTRAINED AT 550°F, TESTED AT -16°F.

BAR	PRESTRAIN	SURFACE PREPARATION	FRACTURE LOAD (lb.)		FRACTURE STRESS $4M/b_oh^2$ (ksi)		BAR SIZE
			1st CRACK	FRACT.	1st CRACK	FRACT.	
B-703	0.33	BARS MACHINED BEFORE PRESTRAIN 0.040 in. EACH SURFACE	-	> 8000	-	> 126	0.710 x 1.000 x 7.375 in.
B-704	0.33		-	> 7500	-	> 120	
B-701	0.37		-	> 8000	-	> 126	
B-702	0.37		-	> 7500	-	> 122	
B-707	0.43		-	> 7500	-	> 129*	
B-708	0.43		-	> 7500	-	> 130*	
B-705	0.43		50	1600	11	-	
B-631	0.46		-	> 7500	-	> 130*	
B-632	0.46		800	1800	19	-	
B-633	0.50		500	1500	11	-	
B-634	0.50		-	2100	-	46	
B-706	0.50		-	3500	-	78	
B-635	0.55		100	1600	2	-	
B-636	0.56		50	1500	1	-	
B-638	0.63		50	1100	1		
B-523	0.24		-	> 6000			
B-524	0.30		-	> 6000			
B-525	0.30		-	> 6000			
B-536	0.30		-	> 6000			
B-537	0.30		-	> 6000			
B-530	0.35		-	> 6000			
B-531	0.35		-	> 6000			
B-543	0.35		-	> 6000			
B-544	0.35		-	> 6000			
B-534	0.38		-	> 6000			
B-535	0.38		-	> 6000			
B-541	0.41		-	> 6000			
B-542	0.41		-	> 6000			
B-545	0.42		-	> 6000			
B-546	0.42		-	> 6000			
B-547	0.45		-	> 6000			
B-548	0.45		-	> 6000			
B-521	0.47		-	> 6000			
B-522	0.47		-	> 6000			
B-518	0.51		50	1100			
B-527	0.52		-	50			
B-519	0.54		50	800			
B-526	0.56		-	50			

* Large plastic strains, 0.22 to 0.24, appearance of stars and tiny shear cracks on the surface.

TABLE VII AR(550)M ABS-B STEEL REVERSED-BEND TEST. BARS MACHINED AFTER PRESTRAINING. PRESTRAINED AT 550°F, TESTED AT -16°F.

BAR	REDUCED PRESTRAIN	MACHINED AFTER PRESTRAIN Δh in.	FRACTURE LOAD (lb.) 1st CRACK	FRACTURE LOAD (lb.) FRACT.	FRACTURE STRESS $4M/b_oh^2$ (ksi) 1st CRACK	FRACTURE STRESS $4M/b_oh^2$ (ksi) FRACT.	BAR SIZE
B-762	0.28	0.014	-	> 7000	-	> 88	0.790 x 1.000 x 7.375 in.
B-763	0.31	0.013	-	> 7000	-	> 93	
B-764	0.36	0.014	-	> 3800	-	> 65*	
B-765	0.36	0.014	-	> 5100	-	> 100	
B-779	0.40	0.017	-	> 7000	-	> 112	
B-780	0.40	0.014	-	> 7000	-	> 112	
B-770	0.41	0.025	-	> 7000	-	> 110	
B-771	0.41	0.028	-	> 7000	-	> 110	
B-768	0.45	0.027	-	> 4800	-	> 90*	
B-769	0.45	0.028	-	> 4000	-	> 76*	
B-772	0.48	0.026	-	> 7000	-	> 102	
B-778	0.49	0.027	50	2000	1	-	
B-767	0.44	0.050	-	> 5100	-	> 94*	
B-773	0.45	0.043	-	> 7000	-	> 98	
B-774	0.49	0.047	50	1500	1	-	
B-775	0.52	0.042	500	1500	10	-	
B-776	0.53	0.043	50	1500	1	-	
B-571+	0.38	~0.025	-	> 6000			
B-572+	0.38	~0.025	-	> 6000			
B-575+	0.47	~0.025	-	1100			
B-576+	0.47	~0.025	-	> 6000			

* Fractures started in surface indentations caused during first stage of bending beyond the prestrained region.

\+ Preliminary tests with bars 0.78 x 1.00 x 8.125 in. from plate 71N.

tests show that a significant difference of the two steels is due to an as-rolled layer about 0.040 in. thick, but not to the interior. The two steels have similar compositions and differ only in the finishing temperature, which is lower and better controlled in ABS-C steel, resulting in

TABLE VIII M(70)M ABS-B STEEL REVERSED-BEND TEST. BARS MACHINED BEFORE AND AFTER PRESTRAINING. PRESTRAINED AT 70°F, AGED AND TESTED AT -16°F.

BAR	PRESTRAIN	MACHINED AFTER PRESTRAIN Δh in.	FRACTURE LOAD (lb.) 1st CRACK	FRACTURE LOAD (lb.) FRACT.	FRACTURE STRESS $4M/b_o h^2$ (ksi) 1st CRACK	FRACTURE STRESS $4M/b_o h^2$ (ksi) FRACT.	BAR SIZE
B-714	0.35	0.041	-	> 6500	-	> 98	0.715x1.000x7.375 in. AFTER MACHINING 0.035 in. EACH FACE
B-713	0.36	0.037	-	> 7500	-	> 100	
B-715	0.38	0.032	-	> 7500	-	> 101	
B-716	0.38	0.032	-	> 7500	-	> 100	
B-717	0.40	0.029	-	> 7500	-	> 105*	
B-718	0.41	0.022	-	> 7500	-	> 105*	
B-727	0.51	0.020	-	> 7500	-	> 112*	
B-728	0.51	0.016	-	> 7500	-	> 112*	
B-730	0.58	0.013	-	> 7500	-	> 113*	
B-729	0.58	0.014	-	6500	-	104*	
B-742	0.59	0.015	-	> 6100	-	> 95*	
B-743	0.61	0.022	-	> 7000	-	> 105*	
B-725	0.61	0.023	100	1000	2	-	
B-726	0.63	0.010	1000	1500	23	-	
B-745	0.68	0.014	100	1000	2	-	
B-744	0.68	0.013	800	1000	18	-	

* Large plastic strains, 0.25 to 0.36; appearance of stars and tiny shear cracks on the surface.

finer grain and a lower Charpy V-notch transition temperature. The difference in properties at the interior, studied by Kapadia and Backofen [13], does not seem to be reflected in the exhaustion limits of the steels with machined surfaces. On the contrary, the non-homogeneous tangential strain-

TABLE IX M(550)M ABS-B STEEL REVERSED-BEND TEST. BARS MACHINED BEFORE AND AFTER PRESTRAINING. PRESTRAINED AT 550°F, TESTED AT -16°F.

BAR	REDUCED PRESTRAIN	MACHINED AFTER PRESTRAIN	FRACTURE LOAD (lb.)		FRACTURE STRESS $4M/b_o h^2$ (ksi)		BAR SIZE
		Δh in.	1st CRACK	FRACT.	1st CRACK	FRACT.	
B-643	0.40	0.017	-	> 7000	-	> 108	0.755 x 1.000 x 7.375 in. AFTER MACHINING 0.015 in. EACH FACE
B-644	0.41	0.008	-	> 7000	-	> 108	
B-645	0.46	0.007	-	> 7500	-	> 116	
B-646	0.46	0.007	-	> 7500	-	> 115	
B-647	0.47	0.024	-	> 7000	-	> 106	
B-654	0.47	0.039	-	> 7500	-	> 109*	
B-648	0.48	0.015	-	> 7000	-	> 100	
B-653	0.49	0.030	-	> 7500	-	> 113*	
B-738	0.51	0.015	-	> 6100	-	> 87	
B-751	0.51	0.011	-	2800	-	60	
B-732	0.52	0.020	-	2100	-	43	
B-734	0.52	0.030	-	3000	-	63	
B-649	0.54	0.007	-	2500	-	51	
B-650	0.54	0.009	-	3000	-	60	
B-737	0.54	0.033	-	4500	-	95	
B-655	0.55	0.020	-	3100	-	65	
B-656	0.55	0.019	-	2200	-	45	
B-739	0.55	0.027	-	2000	-	44	
B-651	0.56	0.041	-	2000	-	34	
B-652	0.56	0.044	-	1800	-	38	
B-749	0.60	0.030	100	1000	2	-	
B-740	0.66	0.026	500	1000	11	-	

* Large plastic strains, 0.25, appearance of stars and tiny shear cracks on the surface.

ing caused by rolling with frictional force, investigated by Hundy and Singer [14], may be closely related with the present findings. The straining was found to be stronger after light than after medium or heavy reduc-

TABLE X M(550)M ABS-B STEEL HOT AXIAL COMPRESSION BARS TESTED IN TENSION AT - 16°F.

BAR	COMPR. TEMP. °F	PRESTRAIN TOTAL	PRESTRAIN STEPS	PRESTRAIN TIME MIN.	0.1% OFFSET STRESS ksi	ULTIMATE STRENGTH ksi	FRACTURE NAT. STRAIN	FRACTURE TRUE STRESS ksi
B-5	530	0.30	4	44	73	100	0.82	141
B-6				44	76	101	0.85	141
B-7				12	68	96	0.88	146
B-8				12	67	97	0.90	145
B-1	527	0.41	4	40	72	106	0.77	137
B-2				7	69	101	0.75	147
B-3				40	71	101	0.83	143
B-4				10	66	98	0.90	155
B-284	550	0.45	5	7	72	100	0.92	158
B-285		0.45	5	7	73	100	0.94	156
B-282		0.49	6	9	72	101	0.90	156
B-283		0.49	6	9	74	103	0.92	164
B-274		0.52	7	12	72	102	0.01	103
B-275		0.52	7	11	72	103	0.87	156
B-280		0.56	8	12	73	97	0.01	97
B-281		0.56	8	13	75	100	0.01*	100*
B-276		0.60	9	15	74	102	0.01	103
B-277		0.60	9	15	73	101	0.01	101
B-278		0.64	10	17	72	93	0.01	94
B-279		0.64	10	17	72	90	0.01	91

* Fracture at Fillet.

tion and to depend on the conditions of surfaces and of rolling.

In laboratory tests of notched or fatigue-cracked plates and in many service failures, fracture appears to start near the plate midthickness and to propagate faster at the interior than at the surface (radial lines from origin, thumbnail, tunneling cracks, shear lip). Nevertheless there is a number of service fractures which have started at the surface close to a weld, where hot straining of the surface layer had certainly occurred. A typical example is the well-known catastrophic fracture of the tanker Ponagansett which originated at the surface adjacent to a welded clip for degaussing cables [15,16]. Other instances of fracture initiation at or close to the surface may be recognized in reference [16-18]. Therefore, it would be of considerable interest to study the specific conditions which cause or reduce the surface damage, such as the temperature of rolling, degree of reduction, heat treatment, composition etc., and also the relation of surface to interior properties. The reversed bend test should prove valuable

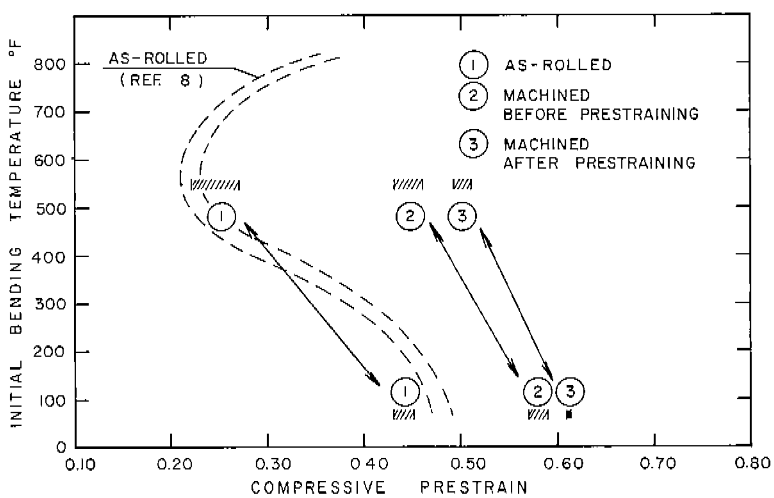


Fig. 10 Exhaustion Limits Of ABS-B Steel Bars With Different Surface Preparations.

in such a study because it offers the almost unique advantage of retaining and testing the initial surface and subjecting it to the highest strain.

In view of the importance of the surface it is interesting to note that some bars machined after hot bending, did not break at the middle of the machined intrados where the nominal compressive prestrain and the reverse bending moment were highest, but at the indentations (Fig. 11) made by the 0.25 in. diameter supports used to load the bar during the first stage of bending (Fig. 1a). Similar indentations were also produced under identical or slightly higher forces on hot bars lying on a continuous flat support. After cooling to -16°F the bars were subjected to bending causing tension on the side of the indentations, but no brittle fracture was obtained. It appears that brittleness resulted not from the indentation strains alone, but from the strain history of rolling and the straining sequence of indentation and bending during the first and especially the second stage of loading (Fig. 1b).

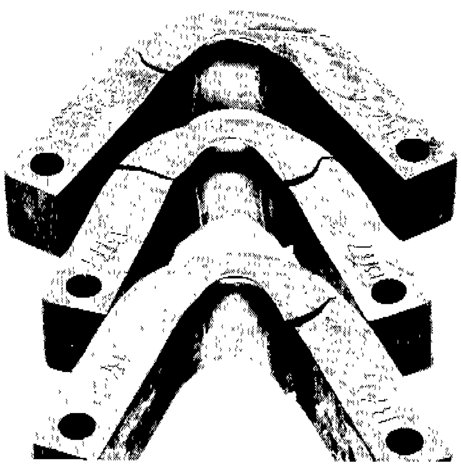


Fig. 11 Fractures Initiated outside Machined Area At Indentations Of As-Rolled Surface Produced During Hot Initial Bending.

All the evidence of a higher surface embrittlement as cause of fracture indicates also that the prestrain damage enhances more the mechanism of fracture initiation than of propagation. Indeed, as shown by the tests with as-rolled surfaces, the interior is already in a state capable of sustaining propagation when the surface strains reach the value of about 0.25 (550°F) or 0.60 (70°F). Nevertheless after surface machining fracture occurs only when the surface strains are increased to about 0.50 (550°F) or 0.60 (70°F). Obviously the more easily embrittled as-rolled layer acts as a trigger or initiator of brittle fracture.

TABLE XI ABS-B STEEL EXHAUSTION LIMITS, FRACTURE STRESSES AND FRACTURE STRAINS MEASURED AT -16°F IN REVERSED-BEND TEST.

Surface Preparation	Test Method	Prestrained At	Exhaustion Limit	Nominal Fracture Stress, ksi*		Nominal Fracture Strain*	
				Brittle	Ductile	Brittle	Ductile
Machined After Prestraining	Reversed-Bend Test	70°F	0.61	2÷23	> 113	0.03÷0.08	> 0.36
		550°F	0.49÷0.51	2÷95	> 116	0.02÷0.08	> 0.25
	Tension	550°F	0.52	103	103 (164+)	0.01	0.61 (max.)
Machined Before Prestraining	Reversed-Bend Test	70°F	0.57-0.59	1÷48	> 115	0.05	0.45 (max.)
		550°F	0.43-0.46	1÷78	> 130	0.03÷0.06	> 0.24
As-Rolled	Reversed-Bend Test	70°F	0.43-0.45	2÷72	> 130	0.04÷0.09	> 0.18
		550°F	0.22-0.27	5÷60	> 90	0.03÷0.06	> 0.08

* Values refer to prestrains close to the exhaustion limit.
+ True Stress.

6. TESTS IN AXIAL COMPRESSION-TENSION

Tests of ABS-B steel prestrained cold in axial compression (70°F), aged, machined into tension specimens and tested in axial tension at -16°F have already been reported [11] and are shown in Table XI for comparison with the reversed bend tests. Hot axial compression of ABS-B steel was now produced with a specially constructed hot compression machine shown in Figs. 12a and b. operating on the same principle as the cold compression machine [4a]. The bars were of 0.75 in. square cross-section and 9.75 in. length in the direction of rolling and had ground surfaces. They were preheated before being inserted in the compression machine. The bars were held diagonally in the V-grooves of 8 in. long guiding dies pressed against each other with a force of 10,000 to 15,000 lbs.so as to prevent any buckling. The dies were heated by 5 groups of heaters each with its own Variac transformer. The temperature at twelve points was continuously monitored on an autographic recorder with automatic switching, and one of the thermocouples activated also the controller for all heaters. The axial load was applied in steps through consecutively longer and larger plungers matching the shortening and expanding specimen. The consecutive length reductions were approximately 15%, 10%, 8%, 6%, 5% and from then on 4% of the

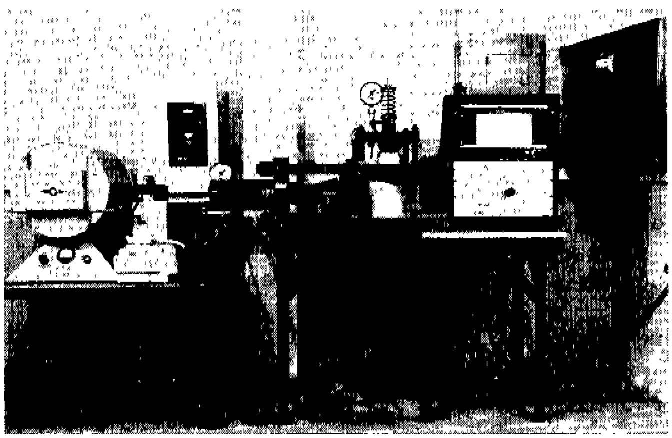


Fig. 12a General View Of Hot Axial Compression Machine. From Left: Preheating Oven, Pump, Oven Temp. Controller, Compression Machine, Autographic Temp. Recorder, Panel With Temp. Controller And Variac Power Controls.

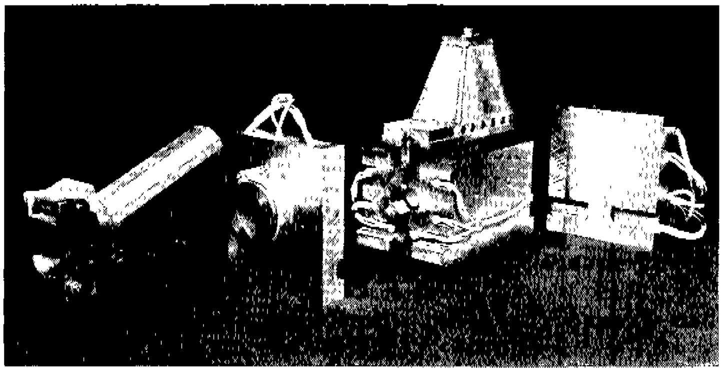


Fig. 12b Exploded View Of Interior Of Hot Compression Machine.

initial length. The duration of each step (loading and an equal time for unloading and changing plunger) was maintained constant for all specimens. After compression the specimens were cooled in boiling water. The total time from the beginning of compression to the moment of cooling was varied from the shortest possible of 3 minutes to 30 minutes in an effort to detect any recovery by heat treatment, but none could be found. Bars of various degrees of hot compression are shown in Fig. 13.

After compression the bars were machined into 0.505 tension specimens and tested in tension at -16°F immersed in a 50% glycerol cooling solution. Load extension diagrams were taken up to the 0.1% offset yield strength. The ultimate strength, the time stress of σ_f and the natural strain ε_f at fracture were found from the fracture load and the final diameter at the neck. Some bars showing ductile or brittle behavior are shown in Fig. 14 The test results are given in Table X and in Fig. 15. The transition from large fracture strains (0.80 - 0.90) to small (0.01) occurs abruptly at the exhaustion limit of 0.52 (Fig. 14 and 15). This transition is accompanied by a drop of true fracture stress from the range of 140-160 ksi for pre-strains below 0.52, to about 90-105 ksi (equal to the ultimate strength) for prestrains above 0.52. The ultimate strength and the 0.1% offset yield stress remain approximately constant at about 90-105 ksi and about 70-75 ksi respectively for all prestrains. Summarized results are given in Table XI. Raising the prestrain temperature from 70°F to 550°F reduces the exhaustion in axial compression-tension from about 0.70 to 0.52, but does not significantly affect the other quantities.

Bars strained hot, whether axially or in bending, give almost identical exhaustion limits (0.52 vs. 0.50) when machined after prestraining. Cold strained bars show some difference (about 0.70 vs. 0.61) but much less than was

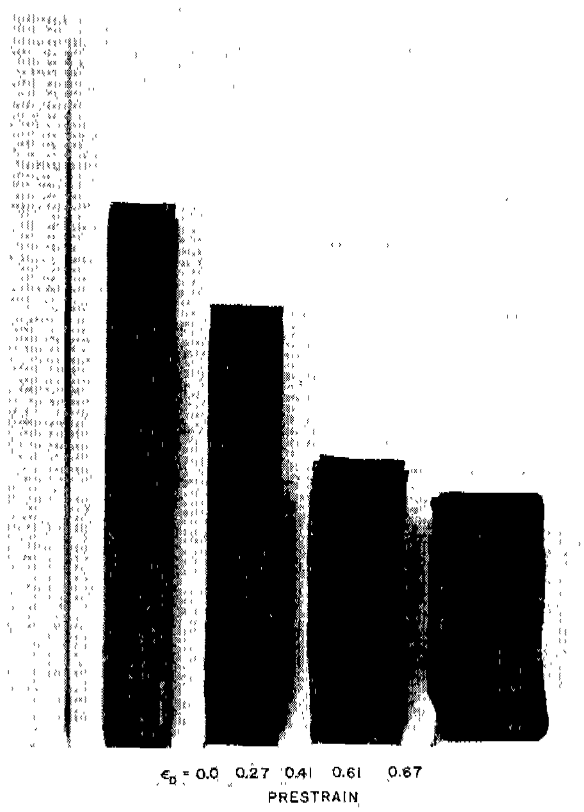
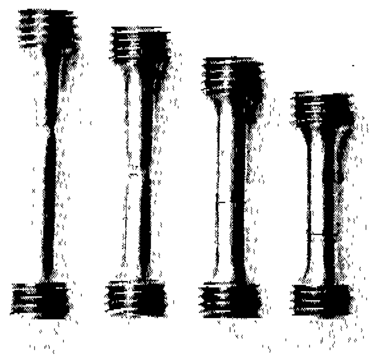


Fig. 13 Bars Axially Compressed At 550°F (Unstrained Bar At Left).



Bar:	B-275	B-283	B-274	B-279
ϵ_o	0.52	0.49	0.52	0.64
ϵ_f	0.87	0.92	0.01	0.01

Fig. 14 Bars Axially Compressed At 550°F Tested in Tension At -16°F.

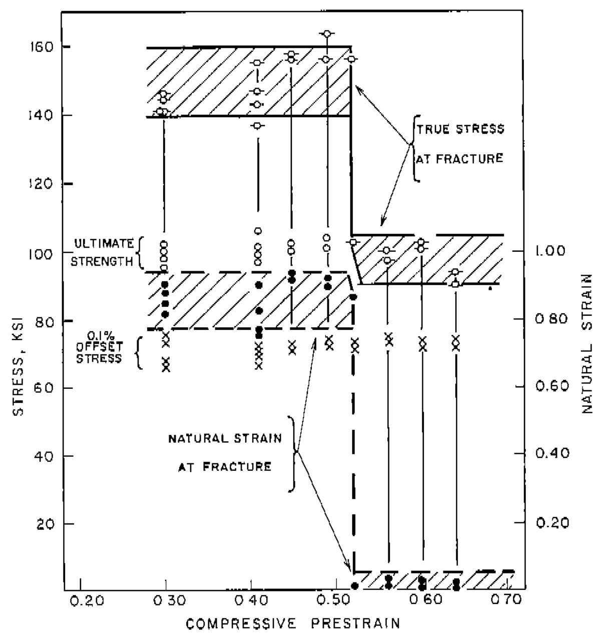


Fig. 15 Transition Of The Mechanical Properties Of Tension Bars Of B-Steel Precompressed At 540°F And Tested At -16°F.

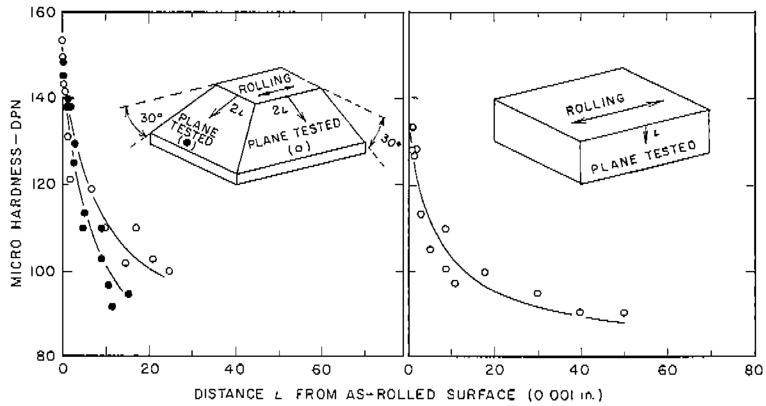


Fig. 16 Micro Hardness Measurements At As-Rolled Surface Of 3/4" Plate No. 77N Of ABS-B Steel.

believed earlier when axial tests (machined) were compared with reverse bend tests of as-rolled bars.

7. STUDY OF THE SURFACE LAYER

The variations of hardness and microstructure of the metal close to the surface confirmed the existence of a harder surface layer. The specimens were thickly plated with chromium as a protection and support of the initial surface against deformation when indentations were made close to it. Microhardness measurements were made on sections perpendicular or at 30° to the inner surface of the bent bars, always close to the plane of symmetry parallel to the plane of bending. A Kentron microhardness tester with 136° diamond pyramid indenter loaded by 10 gramms for 15 seconds was used. Only the ferrite hardness was measured and unreasonably high hardness values were excluded. Nevertheless the scatter was considerable, as is usual with microhardness measurements at such light loads, and averaged values had to be used. The location of the tested plane and the microhardness in terms of depth L from the free surface are given in Figs. 16-20. In an unstrained bar (Fig. 16) the hardness increased toward the edge from about 100 DPN to about 150 DPN on the plane at 30° to the surface and from 90 to about 130 DPN on the normal plane. The increase occurred within a depth of about 0.020 to 0.040 in., as much as the depth of the as-rolled layer which was found to influence the strain embrittlement. A similar bar (as-rolled) pre-strained by 0.34 at 550°(Fig. 17) showed corresponding edge peaks of about 165 DPN on both planes. A bar with a 0.040 in. layer removed before straining by 0.63 at 550°F (Fig. 18) again showed edge peaks of about 140 DPN. Similar hardness peaks were found at the as-rolled edge of a bar cold-strained by 0.45, though slightly smaller than after hot straining (about 140 DPN, Fig. 19, vs. 165 DPN, Fig. 17). Finally bars

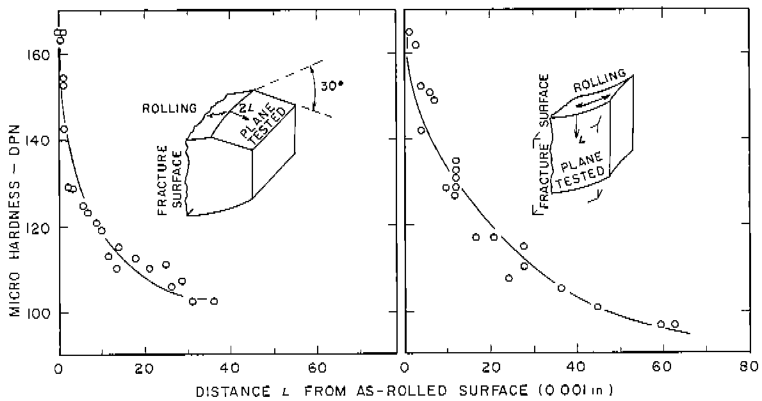


Fig. 17 Micro Hardness Measurements. Bar No. B-710 With As-Rolled Surface, Prestrained To 0.34 At 550°F.

prestrained by 0.37 at 550°F and rendered ductile again by heat treating for 60 minutes at 950°F in an earlier series of tests [8], still showed hardness peaks of about 135 to 140 DPN at the as-rolled surface (Fig. 20). The existence of hardness peaks at the surface of machined bars (Fig. 18) and of heat-treated bars (Fig. 20) raised the suspicion that they might have been caused

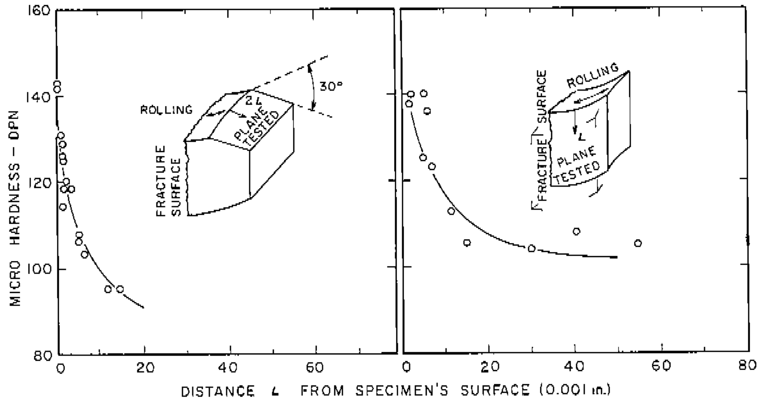


Fig. 18 Microhardness Measurements. Bar No. B-638 With Surfaces Machined Before Prestrain. Prestrained To 0.63 At 550°F.

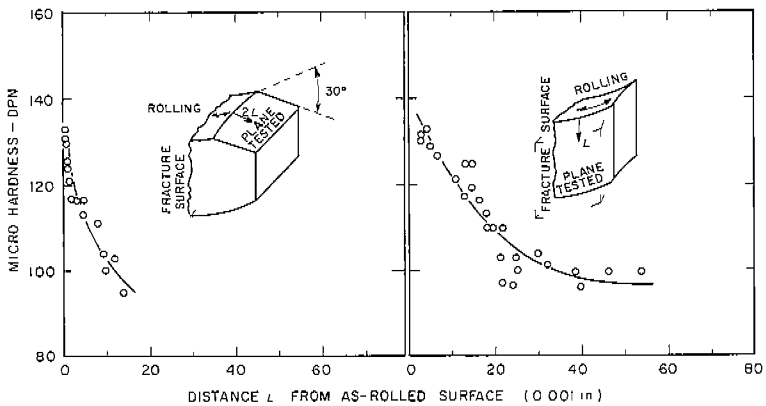


Fig. 19 Micro Hardness Measurements. Bar No. 607 With As-Rolled Surfaces, Prestrained To 0.45 At 70°.

by the chrome plating. This was disproved, however, by measurements close to a plated interior edge of the specimen cut from bar B-384 corresponding to the mid-height of the tested plane in the right hand insert of Fig. 20. No edge peak was found. The unavoidable conclusion must be drawn that prestraining damages and hardens a thin surface layer more than the interior of the specimen. This is confirmed by the tests of bars machined before straining

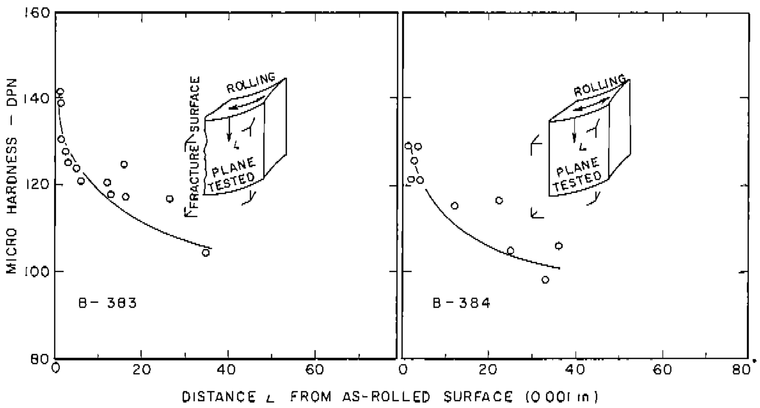
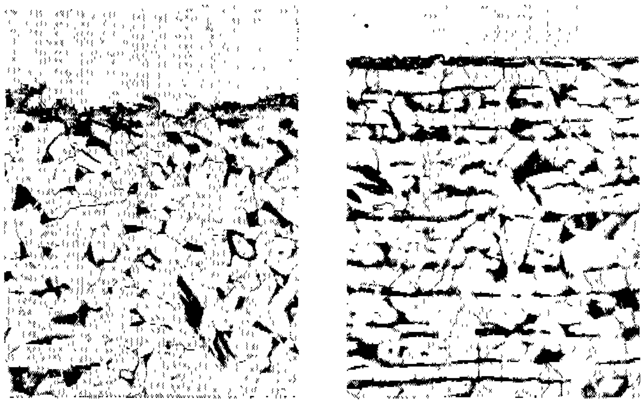


Fig. 20 Micro Hardness Measurements. Bars B-383 And B-384 With As-Rolled Surfaces, Pre-Strained To 0.37 At 550°F And Heat Treated Treated At 950°F For 60 Min.

which show a small but distinct rise of exhaustion limit when the surface is re-machined after straining.

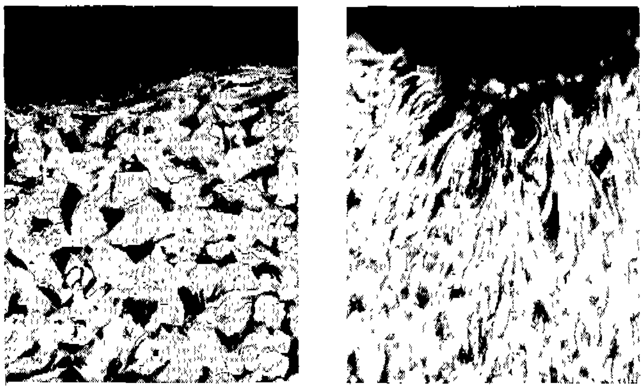
The microstructure of an unstrained bar with as-rolled surface is shown in Fig. 21, to the left on a plane at 30° to the free surface, to the right on a normal plane (the planes indicated in Fig. 16). The deformation close to the surface after hot straining of a machined bar is shown in corresponding planes in Fig. 22, and for cold straining of an as-rolled bar in the normal plane only in Fig. 23. Hot straining obviously causes strong irregularities and inhomogeneous surface strains, apparently even surface folding. In cold straining the irregularity is much less pronounced. The strong straining inhomogeneity of the machined surfaces provides an explanation of their easier embrittlement than the interior. Furthermore the stronger strain inhomogeneity from hot than from cold straining could be the reason for the corresponding easier embrittlement.

The hard and brittle layer which fractures first and triggers the fracture of the ductile bulk of the metal is also found to an extreme degree in bars superficially hardened by nitriding [19]. A similar effect was produced



OBLIQUE NORMAL

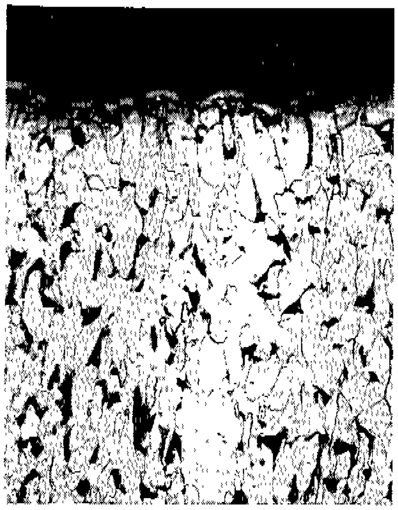
Fig. 21 Microstructure At As-Rolled Surface Of 3/4 in. Plate No. 77N Of ABS-B Steel. Nital Etch, 112X (CF. Gib. 16).



OBLIQUE NORMAL

Fig. 22 Specimen No. B-638, Machined Before Bending. Prestrained 0.63 At 550°F Nital Etch, 110X (CF. Fig. 18).

by a brittle bead welded on the edge of a plate [20], and of course, by the prestrained notch region of the earlier notched plate tests [2,4] which had produced static fractures at a net stress as low as 10% of the initial yield stress. The condition of the as-rolled surface, however, had not been con-



NORMAL

Fig. 23 Specimen No. B-607 With As-Rolled Surface. Prestrained 0.45 At 70°F. Nital Etch, 145X (CF. Fig. 19).

sidered as an important factor in brittle fracture. In welded plates the hardness was found to increase toward the weld and to be higher at 1/8 in. from the faces than at mid-thickness [21].

8. CONCLUSION

A significant factor of brittle fracture of mild steel is its surface conditions, especially when in the as-rolled state. Removal of the as-rolled surface after prestraining resulted in an increase of exhaustion limit (or limit of compressive strain causing brittleness in tension) by about 0.17 after cold straining and by as much as 0.25 after hot straining. The embrittlement by prestraining was found to be higher at the surface than at the interior even when the as-rolled surface had been removed before straining. Microhardness tests showed that the more embrittled layer was also appreciably harder than the interior.

The sensitivity of the surface to embrittlement, especially of the as-rolled surface, is important whenever fracture is likely to be initiated at the surface, as indeed has happened in several service structures. The actual mechanism of embrittlement is not clear, but it is certainly associated with the rolling history and finishing temperature and with the subsequent strain history, especially at high temperatures close to welds in regions of strain concentration. Accordingly it would seem useful to study the influence on surface straining and subsequent embrittlement of such factors as rolling temperature and degree of reduction of grain size and generally of composition. Such studies could indicate the more damaging practices to be avoided and could significantly reduce a cause of brittle fracture.

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The compressive prestrain (exhaustion limit) needed to cause brittle behavior in subsequent tension was found to be much higher in ABS-B steel bars with surfaces machined by about 0.030 in. before straining than with as-rolled surfaces, even more so when the surfaces were machined after straining. Removal of the strained surface caused a small increase of exhaustion limit even when the surfaces had been machined before prestraining. In all cases the increase was larger for bars prestrained at 550°F than at 70°F. The surface effect was found stronger than in earlier tests with an ABS-C steel. In addition the microhardness was found to rise gradually in a 0.030 in. layer adjacent to the surface and to reach a peak at the surface itself in all as-rolled or as-strained surfaces.

The surface damage from an unfavorable rolling history permits an easier surface embrittlement by hot straining in a region of strain concentration close to a weld and creates a dangerous trigger of brittle fracture, as is indicated by service fractures starting at such regions. A study of the rolling and straining history causing such weak regions could help their prevention.

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