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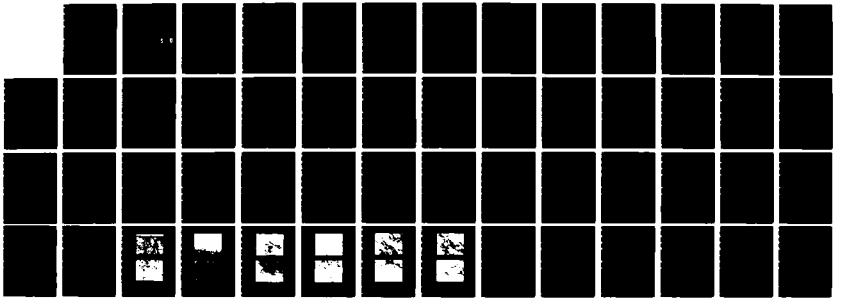

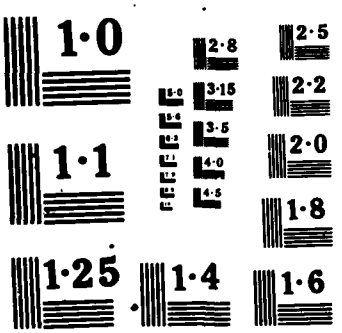

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AMMRC TR 85-25

CARBURIZING STEEL FOR HIGH TEMPERATURE SERVICE

August 1985

T. B. CAMERON and D. E. DIESBURG
AMAX Materials Research Center
1600 Huron Parkway
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FINAL REPORT Contract No: DAAG46-82-C-0066

Prepared for

ARMY MATERIALS AND MECHANICS RESEARCH CENTER
Watertown, Massachusetts 02172-0001

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ABSTRACT

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Carburizing Steel for High Temperature Service

T. B. Cameron and D. E. Diesburg
AMAX Materials Research Center
Ann Arbor, Michigan 48105

Final Technical Report AMMRC TR 85-25
August 1985, 45 pages
Contract DAAG46-82-C-0066
D/A Project: 612105.H840011
Final Report: Sept. 1983 to Aug. 1985

Key Words
1. Steel
2. Gears
3. Alloy Development
4. Heat Treatment
5. High Temperature
6. Carburizing

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Report 82-C-66
February 27, 1985

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T. B. Cameron and D. E. Diesburg

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Key Words: Carburizing steels, CBS1000, M50, temper resistance, hot hardness, fracture toughness, rolling contact fatigue.

INTRODUCTION

A prime concern in airborne equipment is to avoid a brittle fracture in gearing or bearings which could lead to catastrophic engine or propulsion system failure. Projected requirements for advanced aircraft and helicopters suggest that currently employed through-hardened materials will no longer be adequate for these applications due to their low fracture toughness. Hence there is an interest in carburized materials which have somewhat similar surface characteristics to the through-hardened materials but inherently higher core toughness due to the lower core carbon levels. Design specifications indicate a successful carburizing steel for these applications would be one which maintains fracture characteristics similar to carburized SAE 9310 and a minimum surface hardness of 58 HRC after a 1000-hour exposure to 315 C (600 F) temperatures. SAE 9310 has good fracture toughness and hardness at temperatures below 150 C (300 F) but does not maintain adequate hardness at the operating temperatures expected in critical gearing and bearing applications.

Previous research results[1] have shown that CBS1000, a relatively high alloy carburizing steel made by the Timken Company, may have optimum properties in comparison with other available compositions, but its fracture toughness is inferior to that of SAE 9310 and its surface hardness after exposure is only marginally within specifications. Hence, the objective of this investigation was to evaluate various alloy and processing modifications of the CBS1000 base composition which previous research had indicated may improve fracture toughness and surface hardness retention. These modifications focused on the influence of silicon, molybdenum and nickel concentrations as well as austenitizing temperature. Also included for comparison was a low carbon modification of M50 which has shown promise for high temperature bearing applications. The composition and processing combination producing optimum toughness and hardness properties was then tested in rolling contact fatigue.

EXPERIMENTAL PROCEDURE

Sample Preparation

The steels in this investigation were initially prepared as 25 kg (55 lb) induction melted heats. Starting materials were pure metals or ferro-alloy addition agents. Melting was conducted under an inert argon atmosphere and each heat was cast into two 78 mm (3-1/16 in.) diameter ingots approximately 200 mm (8 in.) in length. Chemical analysis was obtained for each heat from a button which was chill cast on the end of the ingot.

The two ingots from each heat were then welded end to end and used as consumable electrodes for vacuum arc remelting. Hence, induction melted ingots were subsequently remelted in a vacuum arc remelting (VAR) process and cast into a water cooled copper chill mold approximately 105 mm (4-1/8 in.) in diameter. Chemical analysis for light elements was obtained from a section of each remelted ingot following the VAR process.

VAR ingots were then heated to 1200 C (2200 F) and forged to 32 mm (1-1/4 in.) diameter bar from which carbon gradient bars and hot hardness samples were subsequently machined. Half of the 32 mm bars were then reheated to 1200 C and forged and hot rolled into 12 mm (1/2 in.) square bar from which Charpy specimens were machined. Additional specimens were cut from the 32 mm diameter bars for use in determining appropriate carburizing and austenitizing parameters.

Following evaluation of heat treating and fracture toughness data, one steel was selected for rolling contact fatigue testing. An additional amount of this steel was machined into cylindrically shaped specimens for rolling contact fatigue tests and one carbon gradient bar. Carburized and heat treated rolling contact specimens were ground and polished prior to testing, resulting in the removal of approximately 0.25 mm (0.010 in.) from the surface of each specimen.

Carburizing and Heat Treating

To determine the appropriate carburizing potential for the steels in this investigation, 1 mm (0.04 in.) thick wafers from each steel were preoxidized, cleaned and carburized at 925 C (1700 F) for 3.5 hours at carbon potentials of 0.8%, 1.0% and 1.15% and quenched in water. Wafers were then analyzed for total carbon concentration. Carburizing time was such that the ratio of core to surface carbon concentration should be above 0.95.

The effect of reaustenitizing temperature (following carburizing) on the temper resistance of the steels in this investigation was evaluated in two stages. In the first stage, 6 mm (0.25 in.) thick samples of Steels A, E and F were carburized and heat treated as shown in Table 1. Hardness values were recorded at various points in the heat treatment and after 5, 50 and 100 hours of an accelerated tempering treatment at 410 C (770 F). In the second stage, 6 mm (0.25 in.) thick samples of Steels A through E were carburized and heat treated in a similar manner to that shown in Table 1 except that reaustenitizing temperatures ranged from 1037 to 1149 C (1900 to 2100 F), as shown in Table 2. Surface hardness values were recorded before and after a 500 hour exposure at 315 C (600 F). In order to determine the effect of the preoxidation step on final hardness, one sample of each steel was not preoxidized prior to carburization but otherwise processed in a manner similar to the other samples in the second stage.

Heat treatment, carburizing and tempering were conducted on unnotched Charpy specimens, hot hardness samples and carbon gradient bars from each steel in this investigation as described in Table 3. Before carburizing Charpy specimens, two opposite sides were coated with carburization inhibiting paint to prevent carburization on those faces.

As-heat treated surface hardness values on all materials except Steel F were above or slightly below the 58 HRC minimum hardness value. In addition,

subsequent analysis of carbon gradient bars indicated that final surface carbon levels were well below furnace set point values. In an effort to determine if the surface hardness was related to surface carbon levels, Steels A through E were recarburized for either 1.5 or 4 hours and treated again as shown in Steps 3 through 8 of Table 3.

Carbon Gradient Analysis

The carbon gradient bars were heat treated along with the test specimens throughout the heat treating program. They were softened by tempering at 540 C (1000 F) for one hour, and chips were machined in incremental layers for carbon analysis by a combustion method.

In order to obtain carbon analysis of recarburized (11 hour) specimens, the side portions (stopped off during initial 7 hour carburizing) of the Charpy specimens were machined off following softening, and ten layers were removed from the remaining carburized surfaces in increments of 0.25 mm (0.010 in.).

Hardness Testing

The surface hardness of the carburized Charpy samples was evaluated before the treatment for 1000 hours at 315 C (600 F) by taking HRA measurements directly on the carburized surface and converting the readings to HRC values. Microhardness profiles were also obtained from representative samples which had been mounted and polished. The hot hardness of each carburized steel was determined under vacuum between room temperature and 400 C (725 F). Vickers hardness impressions were made directly on the carburized surface with a 2.5 kg load on specimens heated in 50°C (90°F) increments starting at 100 C (212 F). The surface of the specimen had been lightly cleaned with 600 grit paper prior to hot hardness testing. The hot hardness of the core of each steel following heat treatment was determined in a similar manner on specimens from which the case had been ground off.

Microhardness profiles for the specimens used in rolling contact fatigue testing were obtained from the carbon gradient bar processed along with the rolling contact specimens. Surface hardness values were obtained on rolling contact specimens following machining and polishing and prior to rolling contact fatigue tests.

Fracture Toughness Testing

Fracture toughness testing was conducted on both carburized (8.5 hour) and recarburized (11 hour) specimens in either the as heat treated condition or following the 1000 hour exposure to 315 C (600 F). Prior to testing, notches 0.13 mm (0.005 in.) wide were machined into the surface of the Charpy bars using electrodischarge machining (EDM) to depths which ranged from 0.13 mm (0.005 in.) to 1.5 mm (0.060 in.). The location of the notches was

the same as that used for normal Charpy V-notch specimens (see ASTM E23 Standard Testing Procedure). The notches were sharpened by high-cycle fatigue precracking between a constant minimum and maximum load, where $P_{max} = 10\ P_{min}$. In general, precracking was successful with values of $P_{max} = 635$ kg (1400 lb); however, loads of up to 726 kg (1600 lb) were needed for the shortest EDM notches. In most cases, successful precracking was obtained in 30,000 cycles.

The precracked specimens were broken in three-point bending as specified in ASTM E399. The load and displacement across the notch opening were recorded and fracture toughness was determined as described for bend specimens in ASTM E399.

Residual Stress Analysis

Following fracture toughness testing, residual stresses were determined as a function of depth from the surface. Sequential layer removal and stress measurement techniques are described elsewhere.[1,2]

Metallography

Fracture toughness specimens were used for metallographic and fractographic examination. Sections were mounted and polished for optical and scanning electron microscope examination of case and core structures. Fracture surfaces were examined directly using the scanning electron microscope.

Rolling Contact Fatigue Tests

Based on a comparison of test results from other aspects of this investigation, Steel D was selected for further evaluation in rolling contact fatigue. Specimens prepared for rolling contact fatigue testing were processed as shown in Table 3 except for the following modifications:

a. Carburizing (Step 2) was conducted for 12 hours.

b. Quenching after reaustenitizing (Step 4) was done in two steps: a flash quench in salt at 620 C (1150 F) followed immediately by a quench in oil at 38 C (100 F).

Rolling contact fatigue tests were conducted on five cylindrical specimens with approximate dimensions of 9.5 mm (0.38 in.) diameter by 76 mm (3 in.) length. The tests were performed by Federal-Mogul using a ball-rod type contact fatigue tester.[3] Final machining and finishing operations on the test specimens were also performed by Federal-Mogul according to their specifications for this test. A minimum of twenty tests was conducted with test lives analyzed by Weibull statistics. Testing conditions are described in more detail in Table 4.

RESULTS

Chemical Analysis

The chemical analysis of each steel following induction melting and of light elements following VAR is given in Table 5. The analyzed compositions were almost identical to the target compositions. Comparison indicates there were only very slight changes in composition as a result of the VAR process. This is as expected, given that the main objective of VAR is to improve chemical and physical homogeneity.[4]

Carbon concentrations obtained in wafers carburized at various carbon potentials are shown in Table 6. From these results, it was concluded that Steels A through E would be carburized at a potential of 1.15%C and that Steel F would be carburized at 0.75%C.

Reaustenitizing Temperature

Stage 1 of the study of the effect of reaustenitizing temperature followed the change in hardness of Steels A, E and F from the carburized and quenched condition through the accelerated (410 C) temper treatment. The average value of hardness for each steel is shown in Figure 1 after the various heat treatment steps described in Table 1. The data indicate that the tempering at 315 C (Step 6) had a substantial effect on the surface hardness, and that no hardness change occurred after the first 5% of the accelerated tempering treatment, Step 9. These results suggest that Steel A is the least temper resistant, that Steel E is marginally below the minimum 58 HRC value and that Steel F easily meets the hardness minimum even at the accelerated (410 C) temperature.

Table 7 presents the results illustrating the effect of the reaustenitizing temperature on hardness of the samples (A, E and F) that were tempered in the accelerated treatment at 410 C. These data also confirm that there was little or no change in hardness after 5 hours at 410 C. These hardness results indicate that the highest reaustenitizing temperature (1095 C) produces the highest values of surface hardness in Steels A and E, but that a temperature of 1040 C produces the highest hardness in Steel F. Although these data indicate it was necessary to use at least a 1095 C reaustenitizing temperature to obtain the minimum 58 HRC following (accelerated) tempering for Steels A and E (and most likely Steels B, C and D also), Steel F had satisfactory hardness results following any of the austenitizing temperatures. Hence, a lower reaustenitizing temperature (950 C/1750 F) was chosen for Steel F.

Surface Hardness and Case Hardness Profiles

As it was not possible to document the precise correspondence of the accelerated 100 hour - 410 C treatment to the required 1000 hour - 315 C treatment, and because of the low and marginal results of Steels A and E, Stage 2

of the study on the effect of reaustenitizing temperature was initiated. This second stage evaluated Steels A through E; Steel F was eliminated because first stage results were considered acceptable. Because maximum tempered hardness values for Steels A and E in Stage 1 were obtained at the maximum austenitizing temperature (1095 C), the range of austenitizing temperatures in Stage 2 was increased to 1150 C as shown in Table 2. The results of the tempering on surface hardness were evaluated after 500 hours at 315 C. Stage 1 results indicated that there was no drop in hardness after the first 5% of the tempering time. Hardness values obtained before and after the Stage 2 tempering treatment are shown in Table 8. These results reinforce those obtained in Stage 1 and indicate there is no change in hardness during extended tempering at 315 C. Except that Steel A has slightly lower hardness following the 500 hour tempering than Steels B through E, average hardness values of the steels before and after tempering are very similar and, in general, slightly above the minimum 58 HRC. A comparison of the results in Tables 7 and 8, showing the influence of reaustenitizing temperature on hardness values, indicates that, for Steels A through E, a reaustenitization at 1095 C (2000 F) should produce the maximum surface hardness values following extended tempering at 315 C. The hardness values of the samples that were not cleaned following preoxidation (Table 7) or not preoxidized at all (Table 8) indicate that preoxidation may not be an essential step prior to carburizing. However, because the influence of preoxidation treatment on case depth and surface carbon concentration has not been investigated, the preoxidation stage was used in this investigation. Based on these results, the heat treatment selected is that shown in Table 3.

The microhardness profiles of Steels A through E carburized for 8.5 hours are shown in Figure 2 along with results of Sample F carburized for 7 hours. Carbon profiles or these same steels are shown in Figure 3. As indicated in Figure 3, surface carbon concentrations are lower than the furnace carbon potential setting. Surface hardness values, however, were considered to be acceptable for the fracture toughness portion of the testing program.

The carbon concentration profiles obtained on the recarburized specimens (11 hours) are shown in Figure 3b. Table 9 shows a comparison of the surface hardness values obtained before and after the 1000 hour exposure at 315 C (600 F) for both the 8.5 hour and 11 hour carburizing conditions. These results confirm that there is very little effect of tempering on the surface hardness at this temperature. The major drop in hardness from the as-carburized value takes place during the two hour 315 C (600 F) temper in Step 6 of the heat treatment program, and there is little subsequent change in hardness.

Results from the hot hardness testing of the carburized cases are shown in Figure 4. The relationship between temperature and surface hardness for Steels A through E is very similar. Over the temperature range up to 200 C (400 F), Steel F has better hot hardness than the other steels. However, beyond this temperature, its hot hardness is similar to the other steels investigated. Hot hardness tests of the core material at temperatures up to 400 C (753 F) are summarized in Table 10. Results indicate that there is little change in core hardness with temperature across this range. Room temperature core hardness of Steels A through D increased with Mo or Si levels

and decreased in Steel E with an increase in Ni level. Core hardness values of Steel F were low due to the formation of carbides and ferrite at the relatively low reaustenitizing temperature employed.

Residual Stress

Residual stress profiles are shown in Figures 5 and 6 for Steels A through F following heat treatment but before the 1000 hour exposure to 315 C (600 F). In general, most values of residual stress were between 0 and -200 MPa (0 and approximately -30 ksi). There was little difference in residual stress between Steels A and B resulting from the difference in carburizing times (8.5 and 11 hours). In addition, Steels A through E had a similar residual stress pattern to Steel F. Previous research[1] indicated that the magnitude of the residual stresses would be reduced by the 1000 hour exposure. Because of the relatively low values obtained before exposure (generally between 0 and -25 ksi), residual stress effects on fracture toughness values were assumed to be negligible following the exposure.

Fracture Toughness

Figure 7 shows a generalized comparison of initial fracture toughness data before and after the 1000 hour exposure at 315 C (600 F). These profiles were not corrected for effects of residual stress. As discussed previously, because of the relatively low magnitude of residual stress values before exposure, the contribution of residual stress to fracture toughness after exposure would be even less and was considered to be negligible. There were 50 specimens evaluated in a pre-exposure condition, and 34 specimens evaluated in a post-exposure condition. The similar results obtained from these two groups of specimens support that there were no significant changes in the fracture toughness values as a result of the 1000 hour exposure.

Based on these results, primary attention was focused on the fracture toughness prior to the 1000 hour exposure. The individual fracture toughness profiles (prior to exposure) are shown in Figure 8. The actual data from which these profiles were obtained are shown in Table 11. These have been corrected for the effects of residual stress. The correction for residual stress was slight, with corrected values generally within 10% of the original value. Steels A through E had relatively similar fracture toughness profiles. The results, however, show a significant difference between the group of Steels A through E and Steel F at carbon levels below 0.7%. The low fracture toughness of Steel F will be addressed in the discussion of microstructures which follows.

Metallography and Fracture Analysis

Metallographic evaluation of the carburized steels was based on a comparison of Steel D and Steel F. Steel D was representative of Steels A-E in

terms of microstructures and fracture appearances. Steel F displayed considerably different microstructures from the other steels in this investigation. Optical micrographs in Figures 9 and 10 illustrate the different structures obtained. Steel D was typical of low alloy carburized steels. However, Steel F had a large amount of carbide present in both the case and core, and the core had a mixed martensite-ferrite structure. The ferrite resulted as an equilibrium phase (rather than a transformation product) from the dual phase austenite plus ferrite matrix structure present at the final austenitizing temperature, 950 C (1750 F). Figures 11 and 12 show similar areas observed at higher magnification in the scanning electron microscope. Figures 13 and 14 illustrate typical fracture surfaces of Steels D and F in the case and core regions of fracture toughness specimens. Though the case fracture appearance is similar in both steels, there were more precipitate particles in evidence (most likely carbide particles) in the fracture surface of Steel F. Core fracture surfaces were dissimilar in that Steel F had a predominantly cleavage type of fracture surface as compared with the mixed dimple-rupture plus quasi-cleavage surface of Steel D.

Rolling Contact Fatigue

Carbon and hardness profiles obtained from the carbon gradient bar processed along with the rolling contact fatigue specimens of Steel D are shown in Figure 15. Note that the specimens themselves would have up to approximately 0.25 mm (0.010 in.) removed from the surface in grinding and polishing prior to testing and that the difference in geometry between the carbon gradient bar and the smaller rolling contact fatigue specimen will result in a slightly deeper case in the latter. Surface hardness values obtained on the ends of the rolling contact specimens are shown in Table 12. Surface hardness values obtained on the rolling contact surface after grinding and polishing but before testing are also shown in Table 12.

Cycle life values obtained in the rolling contact fatigue test are shown in Table 13. Also shown are the results of the Weibull analysis and a 90% confidence band. These results are illustrated in a conventional Weibull plot shown in Figure 16.

DISCUSSION

The objective of this investigation was to focus on steels with compositions similar to CBS1000 (Steels A through E) and look at alloy variations (Si, Mo and Ni) that might improve both temper resistance and fracture toughness. A low carbon version of an M50 composition (Steel F) was also included in the investigation for comparison.

The initial focus of this investigation was the determination of an optimum reaustenitizing temperature, specifically with respect to the surface hardness following the 1000 hour exposure at 315 C (600 F). The results shown

in Tables 7, 8, 9 and 11 indicate that surface hardness values above the 58 HRC minimum can be obtained in the CBS1000 modifications and that these hardness values are maintained during the 1000 hour exposure at 315 C (600 F). Following the selection of Steel D for additional testing, the results in Table 8 indicated that 1095 C would be the optimum reaustenitizing temperature. Results in Table 8, however, suggest that reaustenitizing at 1040 C (1900 F) may be suitable, and commercial practice for CBS1000 suggests that a reaustenitizing temperature of 1010 C (1850 F) may eventually be shown to be acceptable.

Hot hardness results shown in Figure 4 indicate the case hardness characteristics of all the CBS1000 type steels are below those of the M50 steel at temperatures below 200 C (390 F). Steel composition does not appear to play a role in the hot hardness of the CBS1000 types. Core hot hardness results shown in Table 10 illustrate a similar indifference to composition within the CBS1000 steels. The low core hardness value of the M50 steel, Steel F, is a reflection of the large amount of ferrite in the core that resulted from the low reaustenitizing temperature employed.

The carbon profiles shown in Figure 2 suggest that the CBS1000 type steels are relatively resistant to carburizing and that surface carbon levels much above 0.8% are not to be expected following the reaustenitization treatment. The low surface carbon levels of these steels are probably a result of a combination of the relatively high level of nickel in these steels resulting in a retardation of carburization and the high reaustenitization temperature which reduces surface carbon through diffusion. These results are consistent with results obtained in the earlier research on steels with similar nickel levels.[1] Evaluation of the relative effects of Si, Mo and Ni on the carburizing results shown in Table 6 and Figure 3 indicates that Si and Ni additions retard carburization while Mo improves it.

Fracture toughness characteristics of the steels evaluated in this investigation are summarized in Figures 7 and 8. Results from Figure 7 confirm that fracture toughness did not change as a result of the 1000 hour exposure at 315 C (600 F). The results in Figure 8 indicate that composition modifications among the CBS1000 type steels did not influence performance in the fracture toughness tests. A comparison of the fracture toughness profile of Steel D with that of CBS1000 (data from Ref. 1) is shown in Figure 17. The fact that there were differences in the experimental procedures between these two investigations was taken into consideration (see Appendix), resulting in a conservative comparison of the differences between the two steels as shown in Figure 17. The comparison indicates that Steel D has a higher fracture toughness than CBS1000 in the carburized case but less fracture toughness at the low carbon levels representative of the core. However, as explained in the Appendix, the differences between the two steels could be somewhat greater than indicated in Figure 17.

The significant difference in fracture toughness performance between the CBS1000 type steels and the M50 steel was determined to be a result of the ferrite in the core of the M50 steel and the larger carbide volume fraction

in the case of the M50 steel. Hence, a higher reaustenitization temperature for the M50 would likely have resulted in improved core fracture toughness characteristics, and, because it would have reduced carbide volume fraction, an improvement in the case fracture properties would also be realized.

The results of carburizing and testing were reviewed with respect to the selection of one steel for evaluation in rolling contact fatigue. Based on a ranking of Steels A through E with respect to the surface hardness, carbon levels and microstructures obtained, the choice of steels in order of preference was D, C, A, B and E. Steel F had much better surface hardness and temper resistance properties than Steels A through E, but the poor fracture toughness and undesirable microstructure of Steel F suggested that it would not be a suitable choice for further testing. As a result, Steel D was selected for the rolling contact fatigue testing.

The rolling contact fatigue results are shown in Figure 18 in comparison with results obtained on through-hardened M50 material as well as a high alloy carburizing grade, SAE 3310, and a low hardenability grade, SAE 4118. This comparison indicates that Steel D may perform similarly to the through-hardened M50 but not as well as the other carburizing grades. Additional testing with commercially processed material would be required to fully characterize the rolling contact fatigue behavior of Steel D.

CONCLUSIONS

1. Composition variations within the CBS1000 type steels did not have a significant effect on either the temper resistance or the fracture toughness of the carburized case, but they were an important aspect in the ability of these steels to maintain a minimum case hardness exceeding 58 HRC. Of the CBS1000 type steels investigated, Steel D (2.3%Ni) was best suited for maintaining a case hardness exceeding 58 HRC.

2. The carburized low carbon modified M50 steel easily met the minimum surface hardness requirements and exhibited better temper resistance than the CBS1000 type steels at temperatures up to 200 C (390 F). However, the fracture toughness was significantly below that of the CBS1000 type steels.

3. Neither the CBS1000 type steels nor the modified M50 steel exhibited a significant change in fracture toughness as a result of the 1000 hour exposure to 315 C (600 F).

4. The CBS1000 type steel (Steel D) at a hardness level of 58 to 61 HRC had a rolling contact fatigue life similar to that of M50, which had a hardness of 62.5 HRC.

5. The CBS1000 type steel (Steel D) had a higher fracture toughness than CBS1000 in the carburized case but less fracture toughness in the core.

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Table 1

Heat Treatment for Stage 1 of Preliminary Temper Resistance Study (Steels A, E and F)

Step	Procedure
1	Preoxidize at 950 C (1740 F) for 1 hour; glass bead blast oxide off (Sample 6 not cleaned).
2	Carburize at 925 C (1700 F) for 7 hours and warm oil quench. Steels A and E carburized at 1.15%C, Steel F carburized at 0.75%C.
3	Temper at 650 C (1200 F) for 1 hour and air cool.
4	Reaustenitize each steel as shown below in a low dew point hydrogen furnace and warm oil quench.
5	Refrigerate to -80 C (-115 F) for 3 hours.
6	Temper at 315 C (600 F) for 2 hours.
7	Refrigerate to -80 C (-115 F) for 3 hours.
8	Temper at 315 C (600 F) for 2 hours.
9	Temper at 410 C (770 F) for 100 hours.

Sample	1	2	3&6	4	5
Temp. (°C)	950	980	1010	1040	1095
Time (min.)	45	35	25	20	10

Table 2

Reaustenitizing Temperature for Stage 2 of Temper Resistance Study (Steels A-E)

Condition	Temperature, C (F)	Time, min.
1	1040 (1900)	20
2	1095 (2000)	12
3	1150 (2100)	7
4[a]	1095 (2000)	12

[a]Samples in Condition 4 were not preoxidized before carburizing.

Table 3

Heat Treatment Program for Steels A Through F of This Investigation

1. Preoxidize at 950 C (1740 F) for 1 hour.
2. Carburize at 925 C (1700 F) for 7 hours (Steels A-E at 1.15%C and Steel F at 0.75%C) and warm oil quench.
3. Temper at 650 C (1200 F) for 1 hour.
4. Reaustenitize Steels A-E at 1095 C (2000 F) for 10 minutes and Steel F at 950 C (1750 F) for 45 minutes and warm oil quench.
5. Refrigerate to -80 C (-115 F) for 3 hours.
6. Temper at 315 C (600 F) for 2 hours.
7. Refrigerate to -80 C (-115 F) for 3 hours.
8. Temper at 315 C (600 F) for 2 hours.

Note: The above process was modified slightly for the rolling contact fatigue specimens prepared from Steel D as follows:

a. Specimens were carburized for 12 hours.
b. Specimens were reaustenitized at 1095 C (2000 F) for 10 minutes and flash quenched in salt at 620 C (1150 F) immediately prior to oil quenching.

Table 4

Conditions of Rolling Contact Fatigue Tests

Steel	D
Radial Load on Bar	243.6 lb (1084 N)
Calculated Hertzian Stress	786 ksi (5.52 GPa)
Rotating Speed of Bar	3600 rpm
No. Stress Cycles/Bar Revolution	2.389
Lubricant	MIL-L-23699 (Exxon 2380)
Lubricant Temperature	Room temperature (70-75 F)
Lubricant Drip Rate	8-10 drops/minute

Table 5

Composition of Steels Following Various Melting Stages

Steel	Heat	Element, Wt-%											
		C	Mn	Si	Mo	Ni	Cr	V	P	S	Al	N	O
Composition Following Vacuum Induction Melting													
A	P2963	0.12	0.51	0.51	2.28	2.30	1.00	0.41	0.017	0.015	0.060	0.0053	0.0013
B	P2964	0.12	0.51	1.00	2.30	2.34	1.03	0.40	0.016	0.012	0.048	0.0044	0.0015
C	P2965	0.12	0.51	0.51	3.18	2.32	1.01	0.40	0.017	0.012	0.039	0.0051	N.A.
D	P2966	0.12	0.50	0.52	4.22	2.33	1.02	0.41	0.016	0.013	0.055	0.0054	0.0011
E	P2967	0.12	0.50	0.51	3.21	3.04	1.02	0.40	0.017	0.014	0.066	0.0054	N.A.
F	P2968	0.12	0.30	0.25	4.22	NA[a]	4.31	1.31	0.016	0.014	0.020	0.0050	N.A.
Composition Following Vacuum Arc Remelting													
A	P2963	0.13							0.018	0.013	0.051	0.0050	0.0016
B	P2964	0.12							0.017	0.012	0.046	0.0040	0.0023
C	P2965	0.11	Expected to Be						0.018	0.011	0.036	0.0044	N.A.
D	P2966	0.12	Equal to Above						0.018	0.012	0.050	0.0052	0.0020
E	P2967	0.12							0.018	0.013	0.062	0.0048	N.A.
F	P2968	0.12							0.018	0.013	0.015	0.0046	N.A.

[a] NA = not analyzed.

Table 6

Effect of Carburizing Potential on Surface Carbon Content

Steel	Furnace Carbon Potential 0.8%	1.0%	1.15%
A	0.83	0.92	1.04
B	0.67	0.73	1.03
C	0.79	0.90	0.95
D	0.80	0.91	1.02
E	0.80	0.90	1.01
F	1.16	1.43	1.36

Table 7

Surface Hardness (HRC) after Stage 1 Accelerated Tempering Studies

Sample	Reaustenitizing Temp., C (F)	Steel A		Steel E		Steel F	
		Avg.[a]	Final[b]	Avg.	Final	Avg.	Final
1	950 (1750)	52	52	57	55	60	59
2	980 (1800)	54	55	56	57	61	61
3	1010 (1850)	57	57	57	58	61	61
4	1040 (1900)	55	55	55	58	61	62
5	1095 (2000)	58	58	58	60	60	61
6[c]	1010 (1850)	57	57	58	58	61	60

[a]Average of readings taken after 5 hours.

[b]Average value of readings taken after 100 hours.

[c]Sample similar to No. 3 but preoxidized in furnace just prior to carburizing.

Table 8

Surface Hardness[a] of Stage 2 Samples Obtained
Before and After 500 Hours at 315 C

Reaustenitizing Temp., C (F)	Hardness Before Temper, HRC						Hardness After Temper, HRC					
	A	B	C	D	E	Avg.	A	B	C	D	E	Avg.
1040 (1900)	59	58	59	59	60	59	58	57	60	59	60	59
1095 (2000)	59	60	59	56	59	59	58	60	59	60	59	59
1150 (2100)	59	57	59	58	58	58	56	60	59	59	58	58
1095 (2000) Not Preoxidized	58	57	57	60	59	58	56	60	59	60	59	59
Average	59	58	59	58	59		57	59	59	60	59	

[a]Converted from HRA.

Table 9

Surface Hardness (HRC)[a] Before and After 1000 Hours at 315 C (600 F)

Steel	Carburized 8.5 Hours		Carburized 11 Hours	
	Before	After	Before	After
A	57	58	57	57
B	59	58	60	60
C	57	57	58	58
D	59	58	58	59
E	58	56	57	57
F	61	61	--	--

[a]Converted from HRA.

Table 10

Core Hot Hardness Values

Sample	Hardness, HV 2.5 kg	
	Room Temp.	400 C (753 F)
A-1	470	463
B-1	492	461
C-1	503	465
D-1	531	456
E-1	478	480
F	203	182

Table 11

Fracture Toughness at Various Carbon Levels in Carburized Cases (Corrected for Residual Stress Effects

Steel	% Carbon	K_{Ic}[a] MPa$\sqrt{m}$ (ksi$\sqrt{in.}$)
A	0.12	79.5 (72.4)
	0.20	81.6 (74.3)
	0.37	78.2 (71.2)
	0.56	68.2 (62.1)
	0.59	70.8 (64.5)
	0.62	41.0 (37.3)
	0.67	47.7 (43.4)
	0.76	49.0 (44.6)
	0.78	43.8 (39.9)
B	0.12	74.0 (67.4)
	0.30	84.9 (77.3)
	0.44	87.3 (79.5)
	0.53	63.9 (58.2)
	0.54	56.2 (51.2)
	0.58	49.0 (44.6)
	0.58	38.4 (35.0)
	0.60	34.0 (31.0)
	0.61	38.2 (34.8)
	0.65	23.6 (21.5)
	0.71	33.2 (30.2)
	0.72	40.4 (36.8)
C	0.12	85.9 (78.2)
	0.44	64.1 (58.4)
	0.59	38.1 (34.7)
	0.61	48.2 (43.9)
	0.62	45.7 (41.6)
	0.63	34.7 (31.6)
	0.80	36.6 (33.3)
	0.82	46.2 (42.1)
D	0.12	74.2 (67.6)
	0.52	65.4 (59.6)
	0.63	53.1 (48.4)
	0.69	41.1 (37.4)
	0.70	39.1 (35.6)
	0.71	34.0 (31.0)
E	0.12	73.8 (67.2)
	0.42	58.1 (52.9)
	0.56	62.9 (57.3)
	0.61	49.0 (44.6)
	0.62	35.4 (32.2)
	0.62	27.9 (25.4)
	0.69	43.5 (39.6)
	0.75	34.5 (31.4)
	0.76	44.7 (40.7)
F	0.17	41.1 (37.4)
	0.55	25.5 (23.2)
	0.87	22.4 (20.4)
	1.18	18.7 (17.0)
	1.21	13.7 (12.5)
	1.22	26.0 (23.7)

[a] K_{Ic} determined using specimens with short crack lengths.

Table 12

Average Surface Hardness of Rolling Contact Specimens (Steel D)

	HRC
After Heat Treating	59
After Surface Grinding and Polishing	60[a]

[a]Converted from superficial 15-N.

Table 13

Results of Rolling Contact Fatigue Test of Steel D

Item	Hours	Cycles ($\times 10^6$)	Rank %	Item	Hours	Cycles ($\times 10^6$)	Rank %
1	4.50	2.25	3.25	12	13.10	6.55	54.68
2	4.80	2.40	7.92	13	13.30	6.65	59.35
3	7.30	3.65	12.60	14	13.80	6.90	64.03
4	8.00	4.00	17.27	15	14.00	7.00	68.70
5	8.20	4.10	21.95	16	14.60	7.30	73.38
6	9.50	4.75	26.62	17	15.90	7.95	78.05
7	9.70	4.85	31.30	18	16.50	8.25	82.73
8	10.00	5.00	35.97	19	19.60	9.80	87.40
9	11.00	5.50	40.65	20	20.60	10.30	92.08
10	11.30	5.65	45.32	21	33.40	16.70	96.75
11	13.10	6.55	50.00				

21 Items on Test
L10 = 5.86 Hours
2.93 Cycles ($\times 10^6$)

0 Suspensions
L50 = 12.71 Hours
6.35 Cycles ($\times 10^6$)

Slope = 2.45
Correlation = 0.97

90.0% Confidence Band for 21 Completed Tests at Given Slope

Life Level	Low, Cycles ($\times 10^6$)	High, Cycles ($\times 10^6$)
L10	1.90	4.91
L20	2.95	5.71
L30	3.79	6.35
L40	4.55	6.94
L50	5.28	7.50

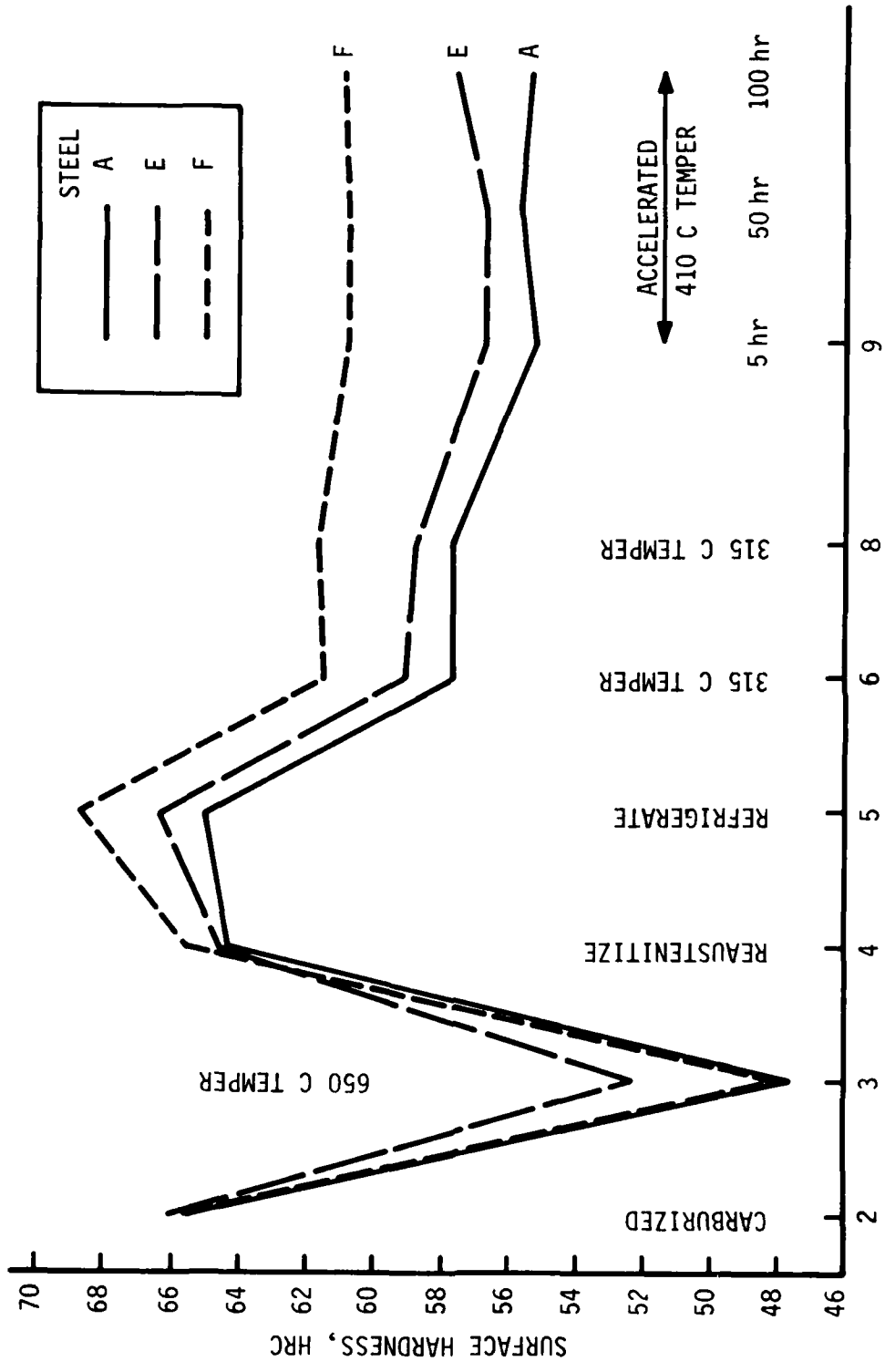


Figure 1 Average Hardness Following Stage 1 Heat Treatments (Described in Table 1)

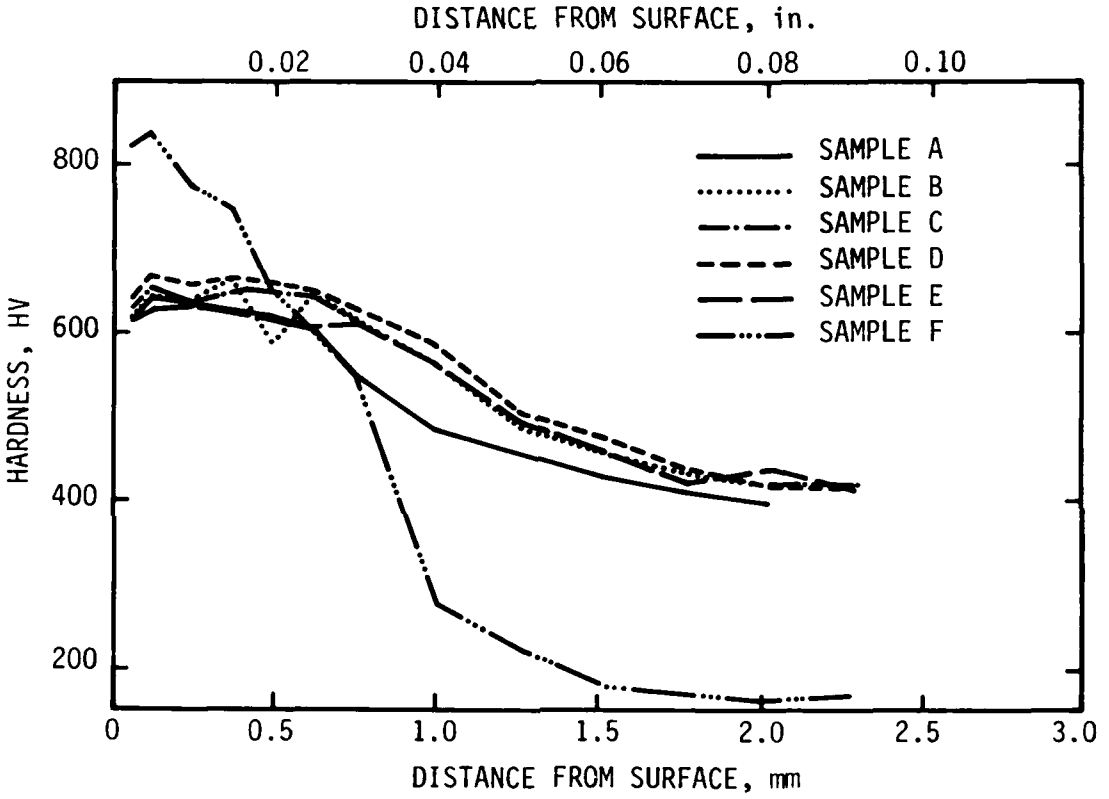
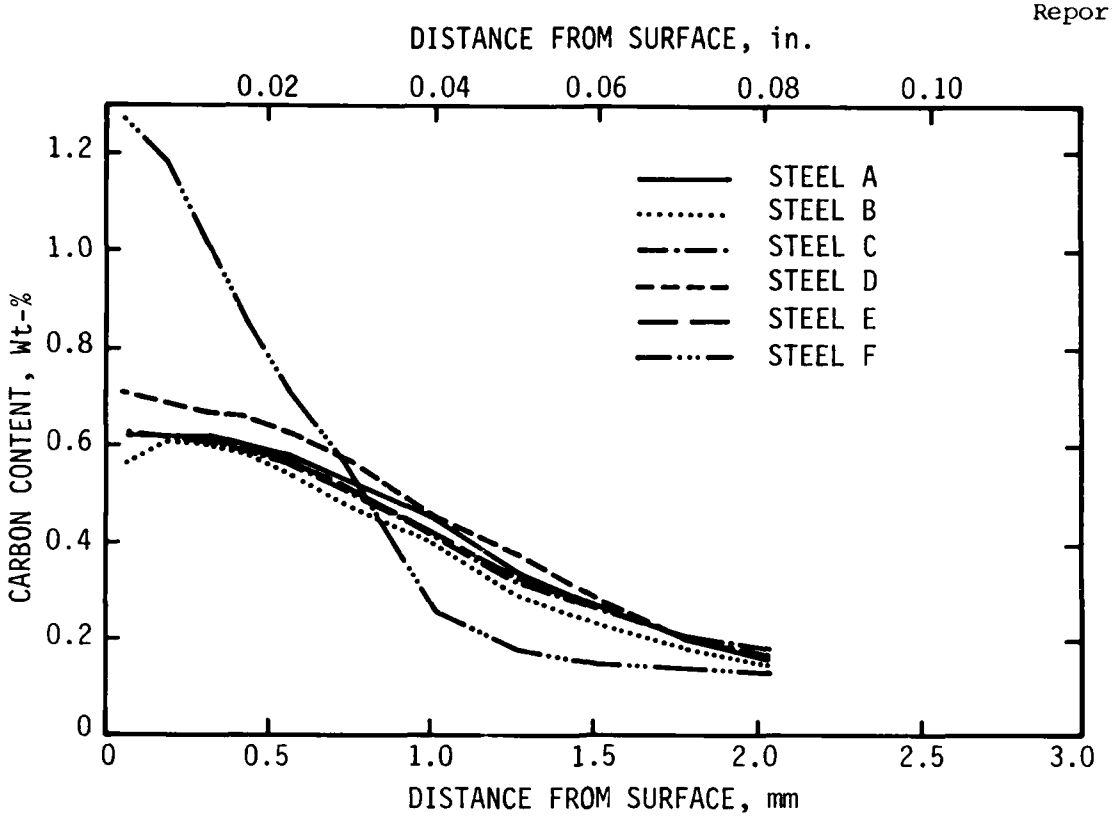
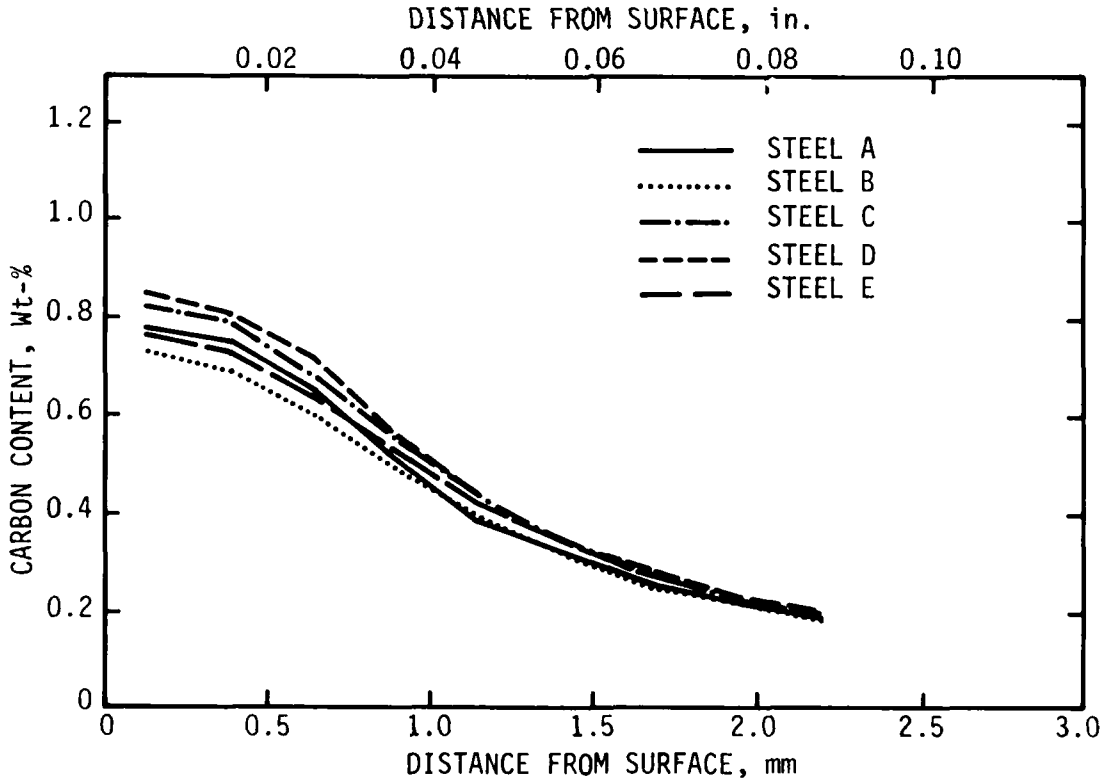


Figure 2 Hardness as a Function of Depth from the Surface for Steels in This Investigation. Steels A-E carburized for 8.5 hours; Steel F carburized for 7 hours.



(a) Steels A - E Carburized 8.5 Hours;
Steel F Carburized 7 Hours



(b) Recarburized Steels (11 Hours Total)

Figure 3 Carbon Content as a Function of Depth from the Surface

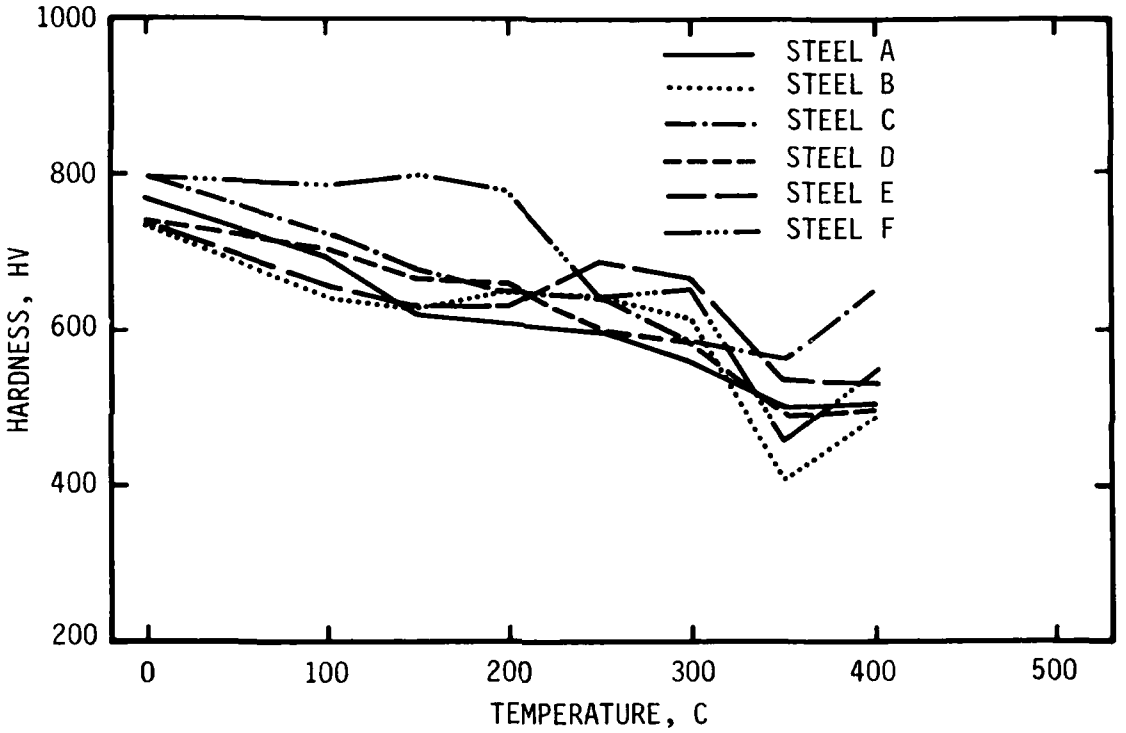


Figure 4 Hot Hardness of Carburized Steels

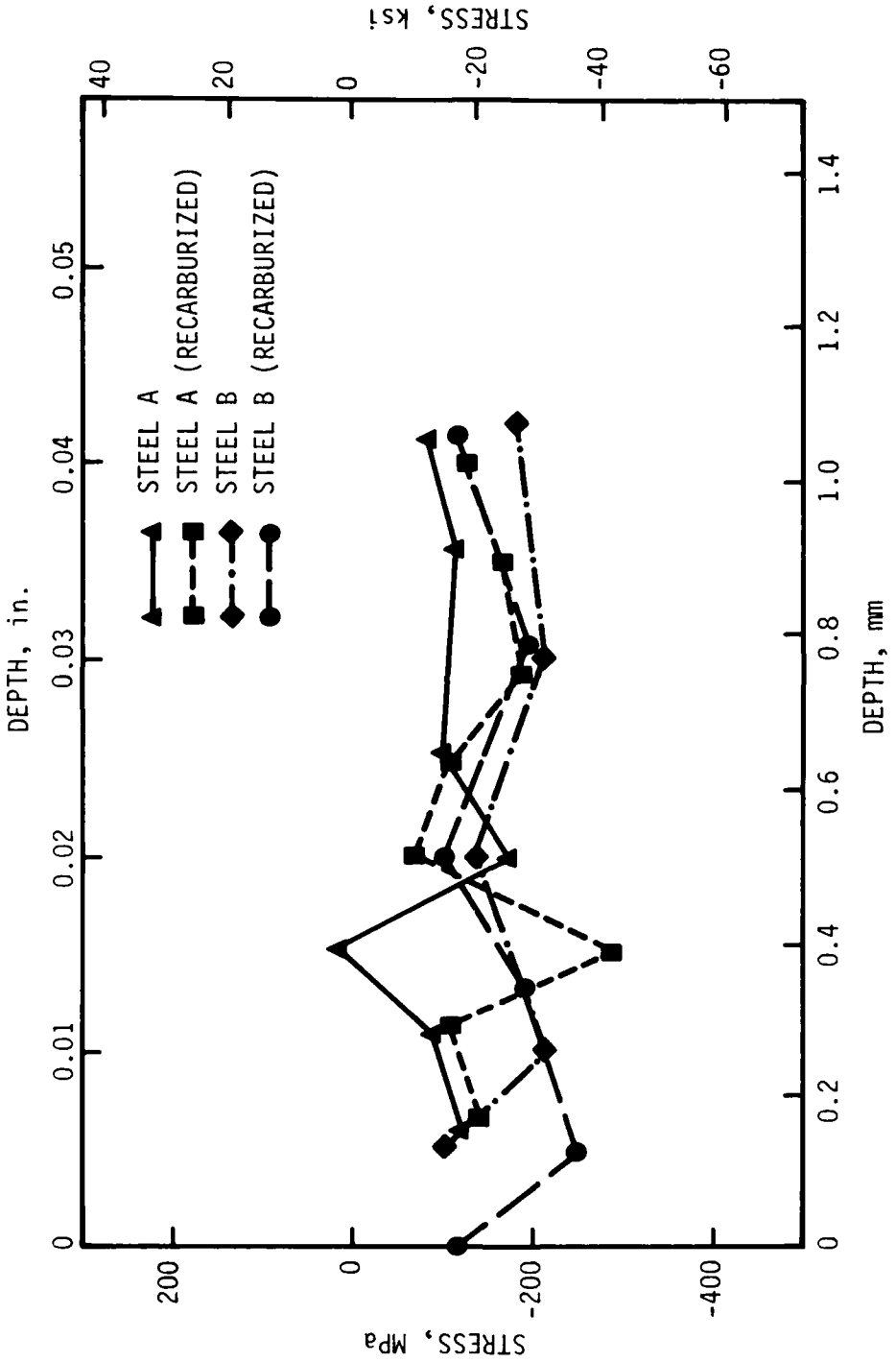


Figure 5 Residual Stress Profiles in Carburized Steels Following Tempering but Before the 1000 Hour Heat Treatment at 315 C (600 F)

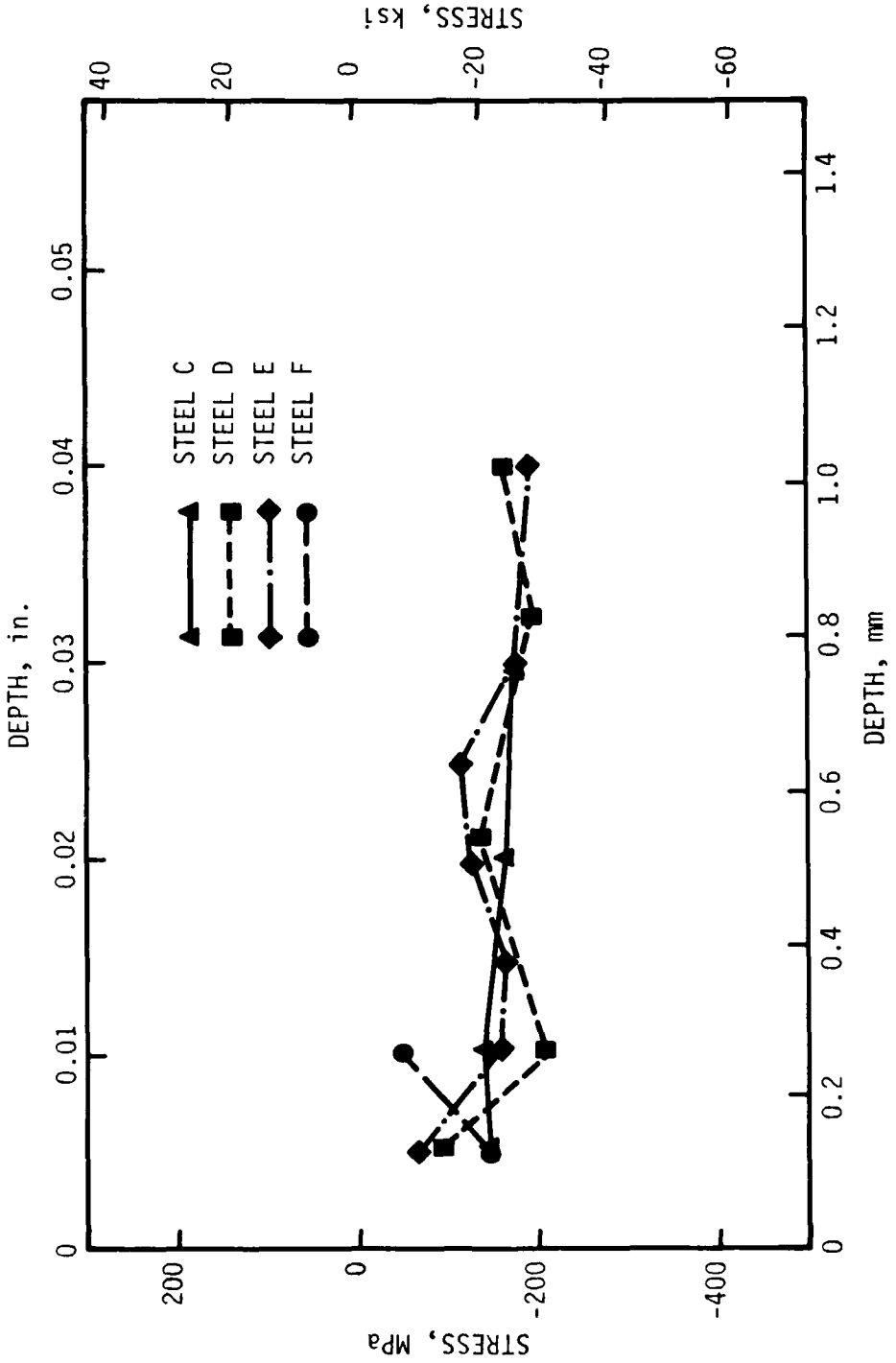
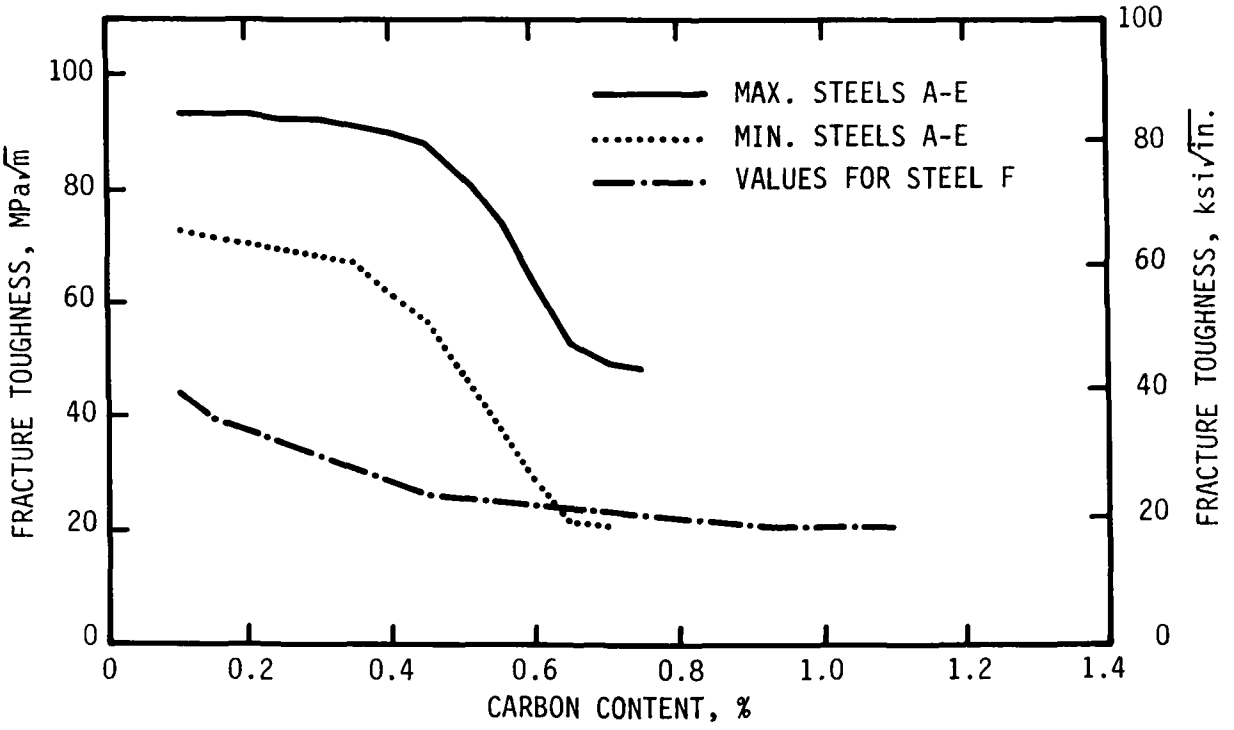
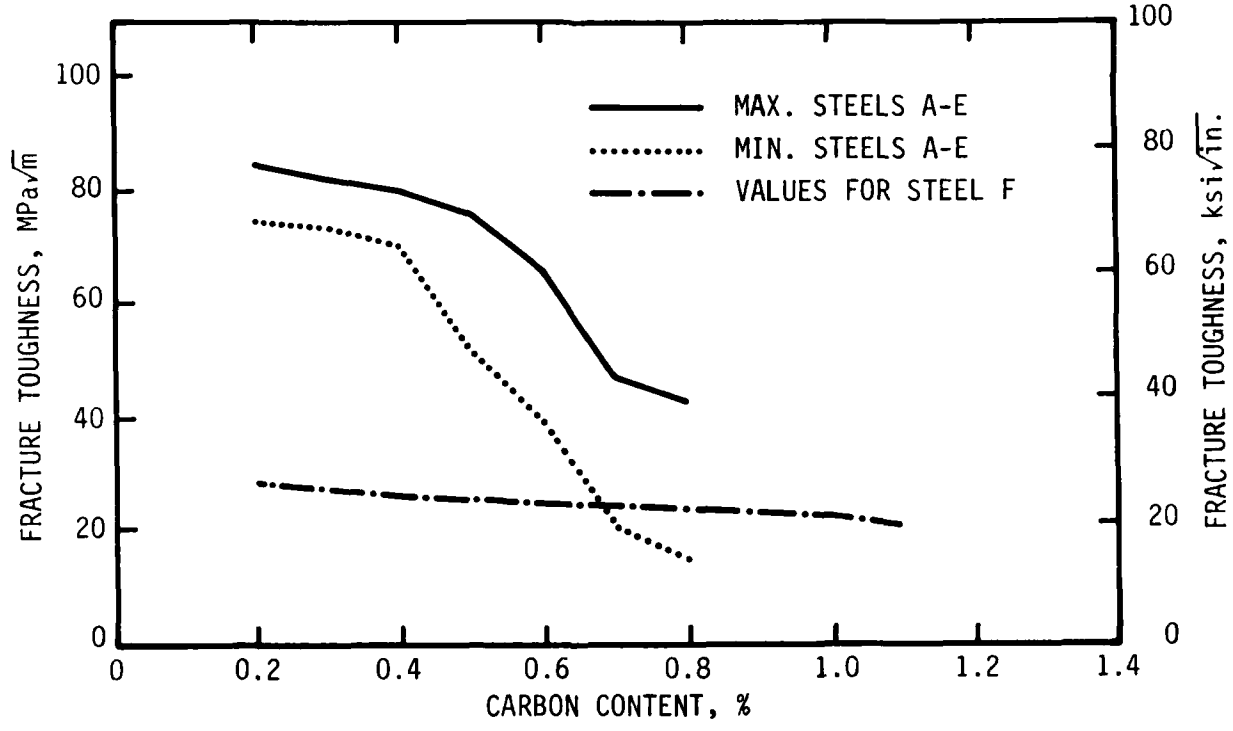


Figure 6 Residual Stress Profiles in Carburized Steels Following Tempering but Before the 1000 Hour Heat Treatment at 315 C (600 F)



(a) Uncorrected Data for Steels Before the 1000 Hour Exposure



(b) Uncorrected Data for Steels Following the 1000 Hour Exposure

Figure 7 Summary of Fracture Toughness Data (Not corrected for the effect of residual stress)

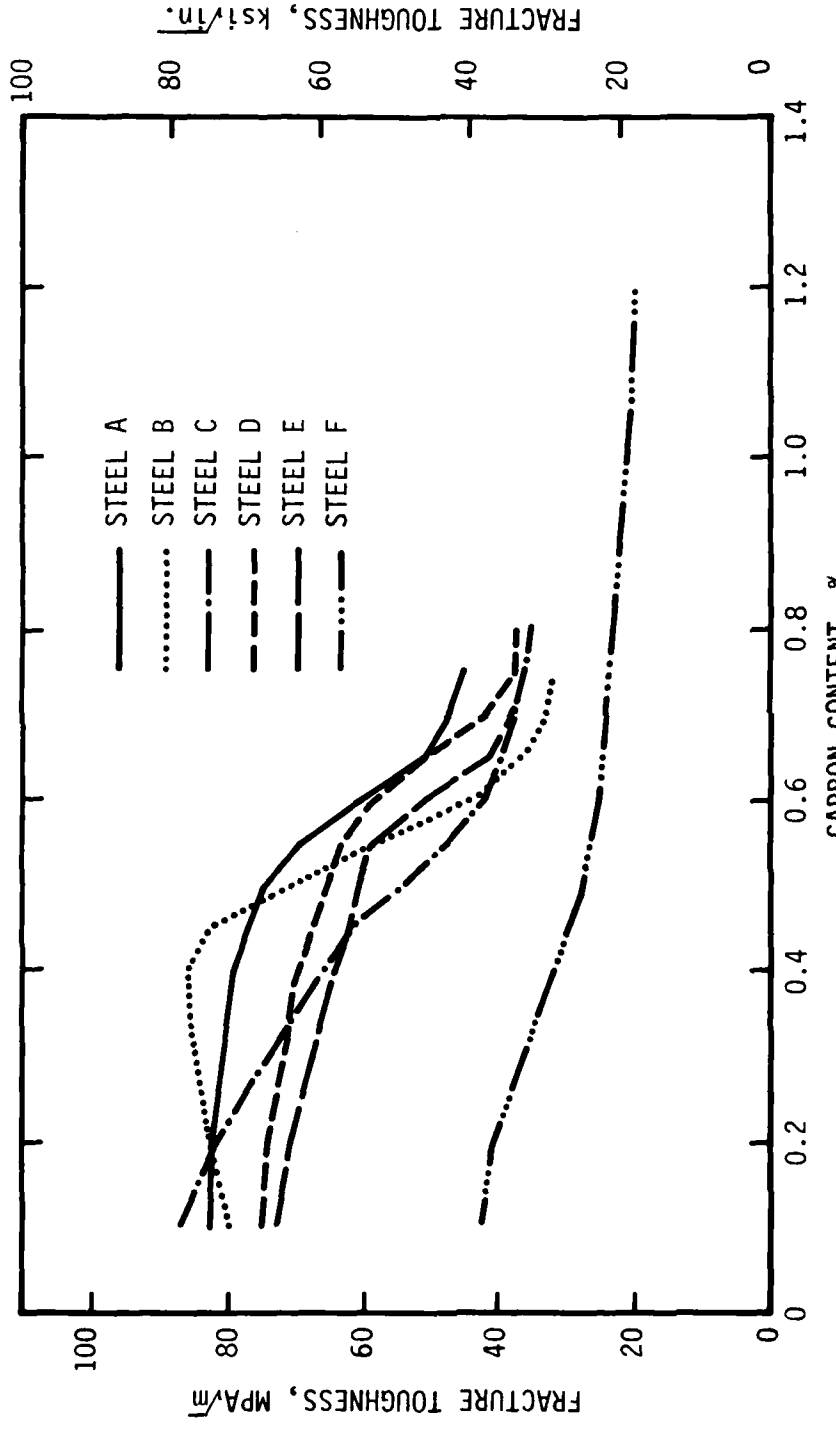
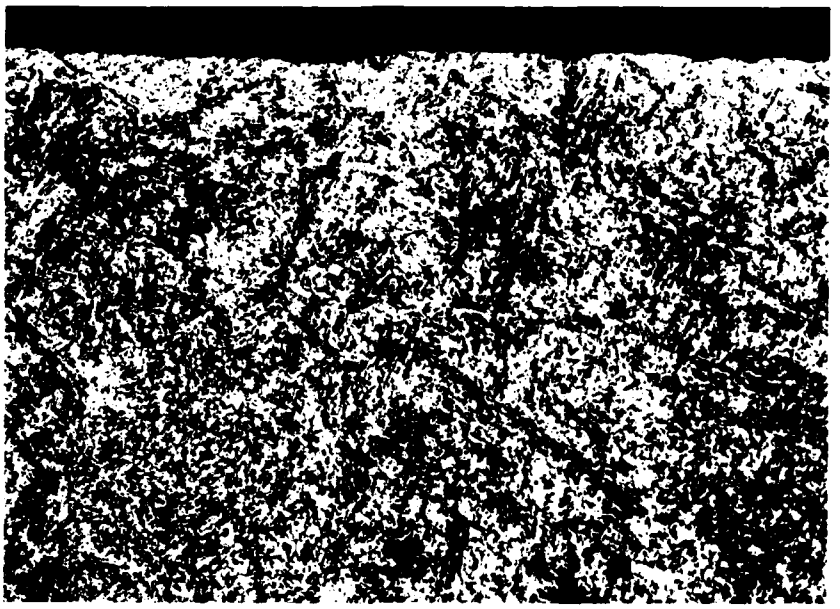
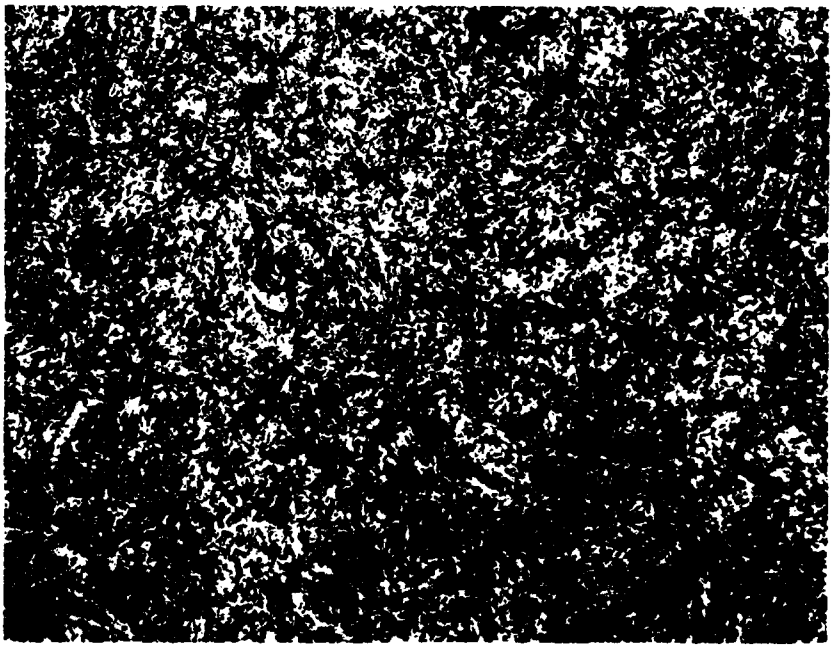


Figure 8 Fracture Toughness Profiles for Carburized Steels Before 1000 Hour Exposure (Corrected for residual stress)



84-559 X500

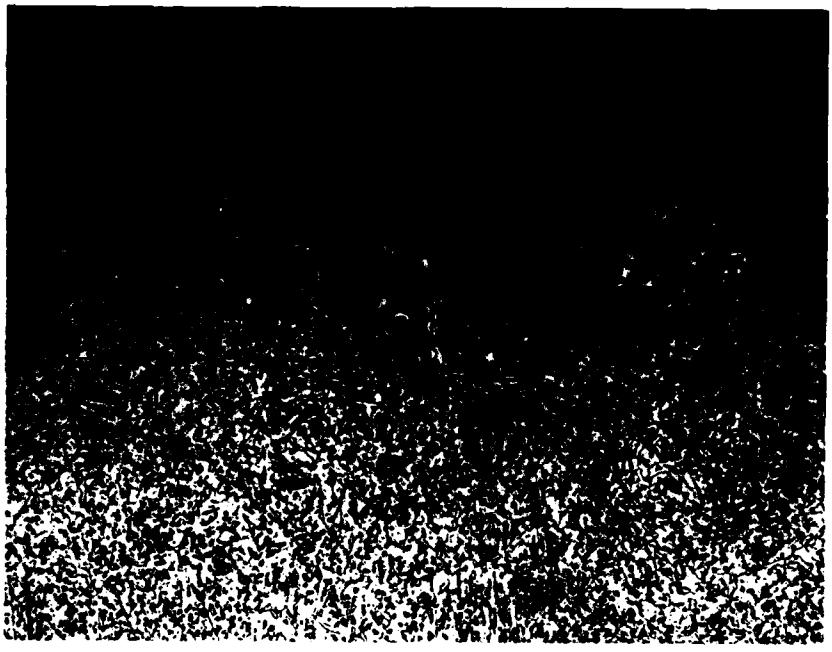
(a) Surface



84-560 X500

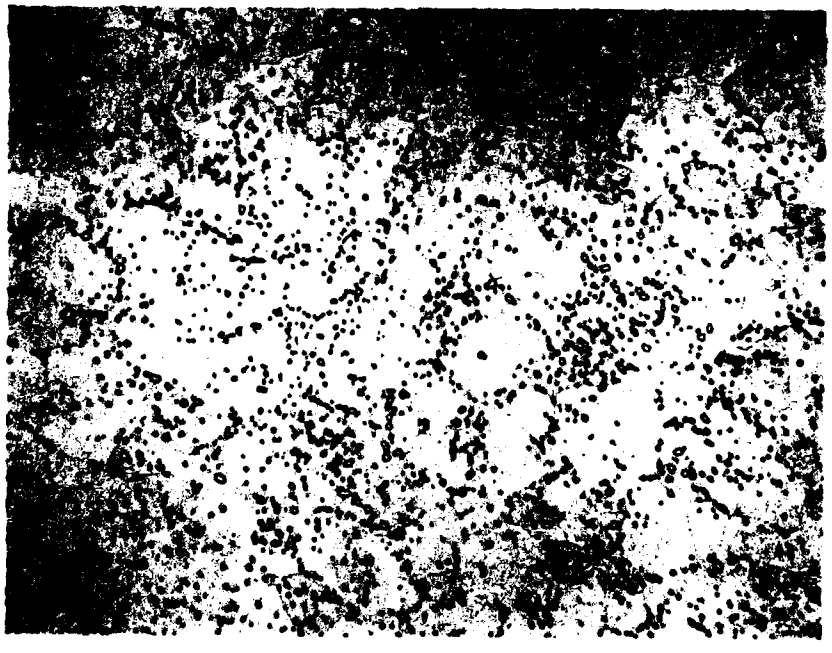
(b) Core

Figure 9 Optical Micrographs from Case (a) and Core (b) Regions of Steel D. Etchant: Nital.



84-562 X500

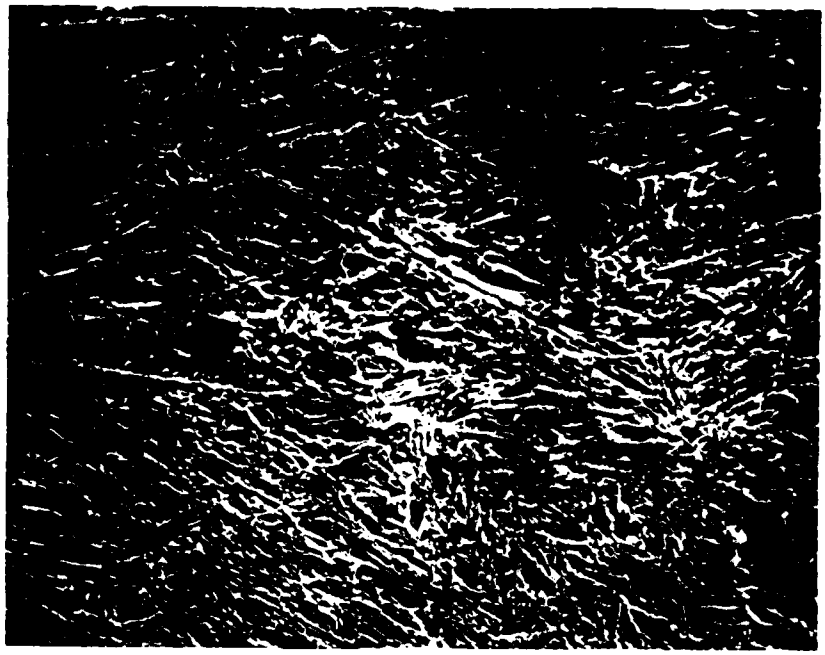
(a) Surface



84-563 X500

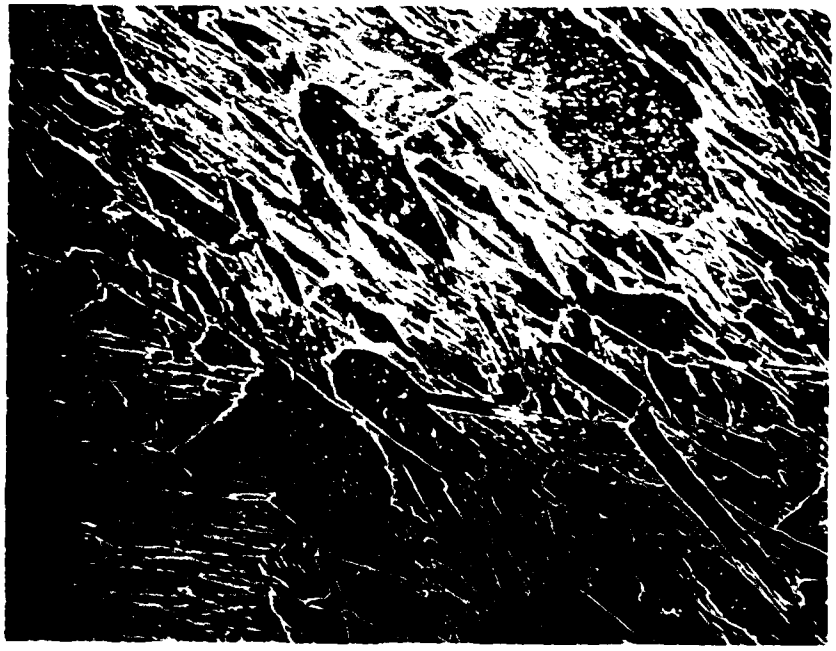
(b) Core

Figure 10 Optical Micrographs from Case (a) and Core (b) of Steel F (Low Carbon M50). Etchant: Nital.



S-8228 X5000

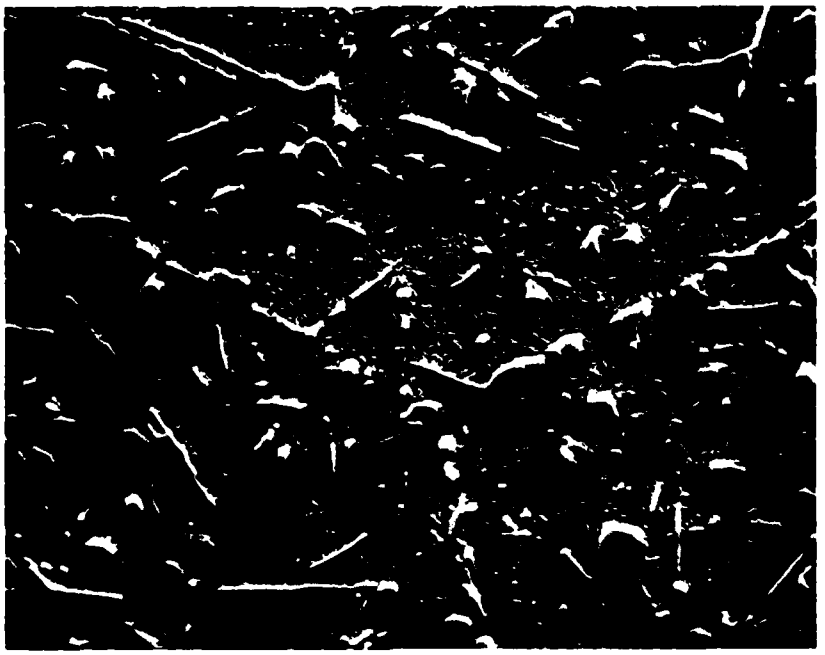
(a) Case



S-8229 X5000

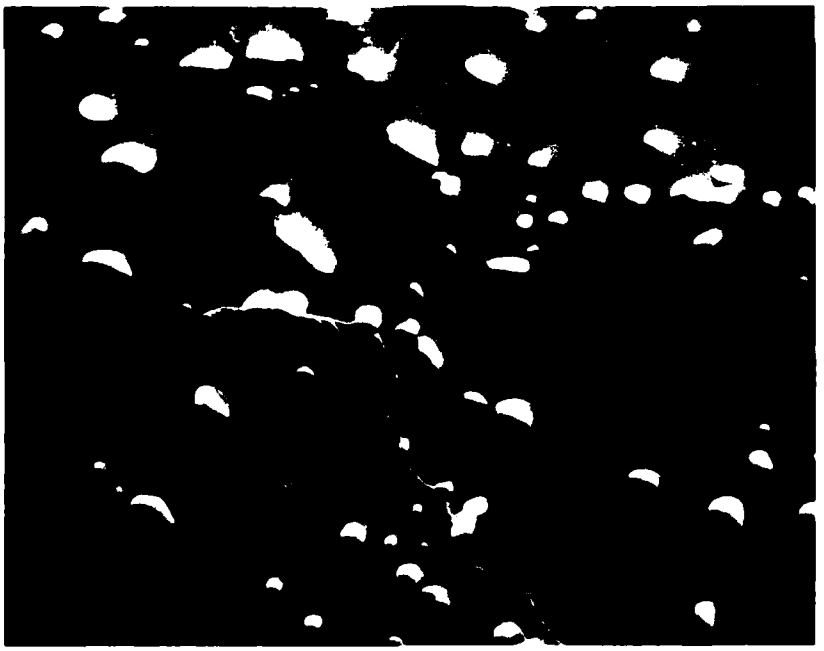
(b) Core

Figure 11 Scanning Electron Micrographs of Case (a) and Core (b) of Steel D Prior to 1000 Hour Exposure. Etchant: Nital.



S-8230 X5000

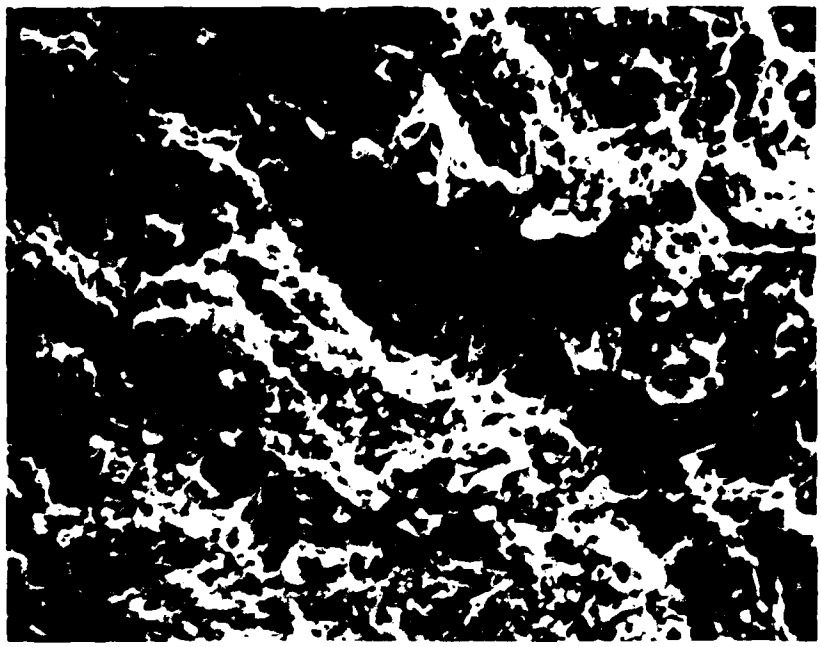
(a) Case



S-8231 X5000

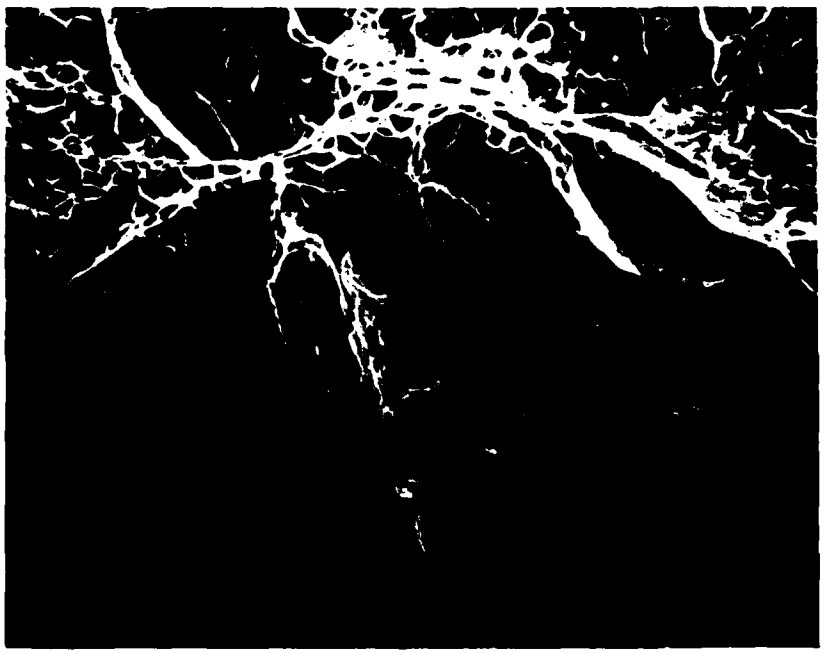
(b) Core

Figure 12 Scanning Electron Micrographs of Case (a) and Core (b) of Steel F (Low Carbon M50) Prior to 1000 Hour Exposure. Etchant: Nital.



S-8236 X2000

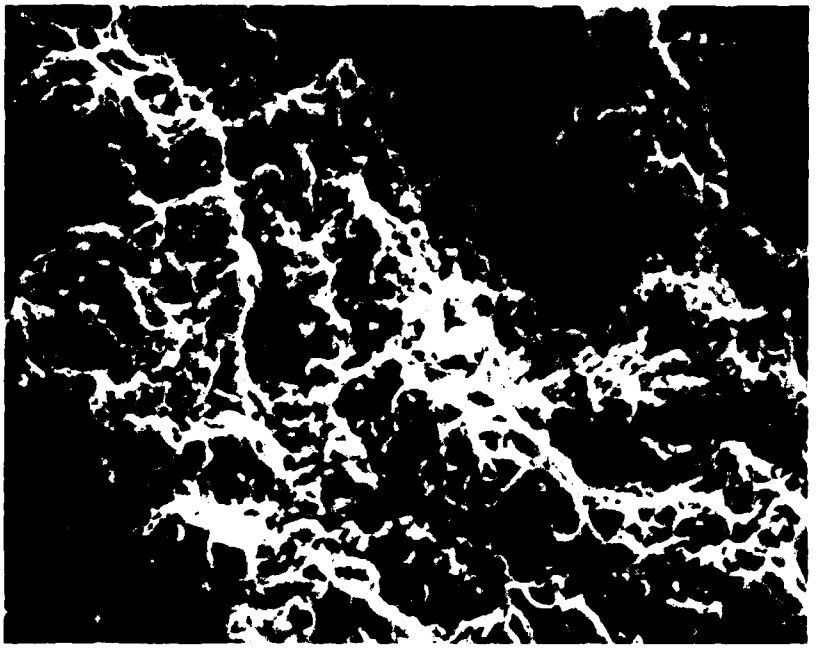
(a) Case



S-8235 X1000

(b) Core

Figure 13 Scanning Electron Fractographs from Case (a) and Core (b) of Steel D Prior to 1000 Hour Exposure



S-8238 X2000

(a) Case

S-8239 X1000

(b) Core

Figure 14 Scanning Electron Fractographs from Case (a) and Core (b) of Steel F (Low Carbon M50) Prior to 1000 Hour Exposure

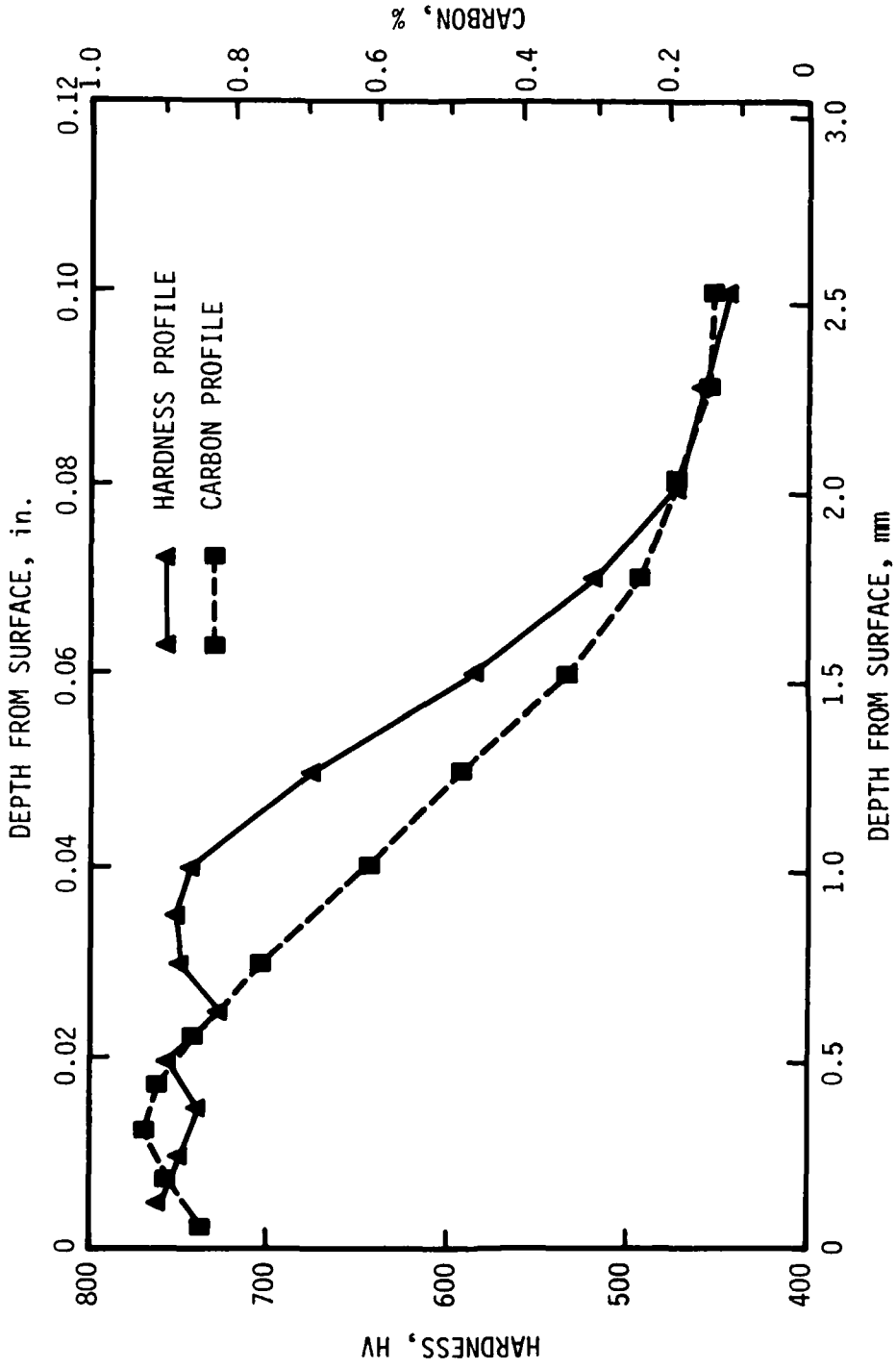


Figure 15 Microhardness and Carbon Profiles in Rolling Contact Fatigue Specimens

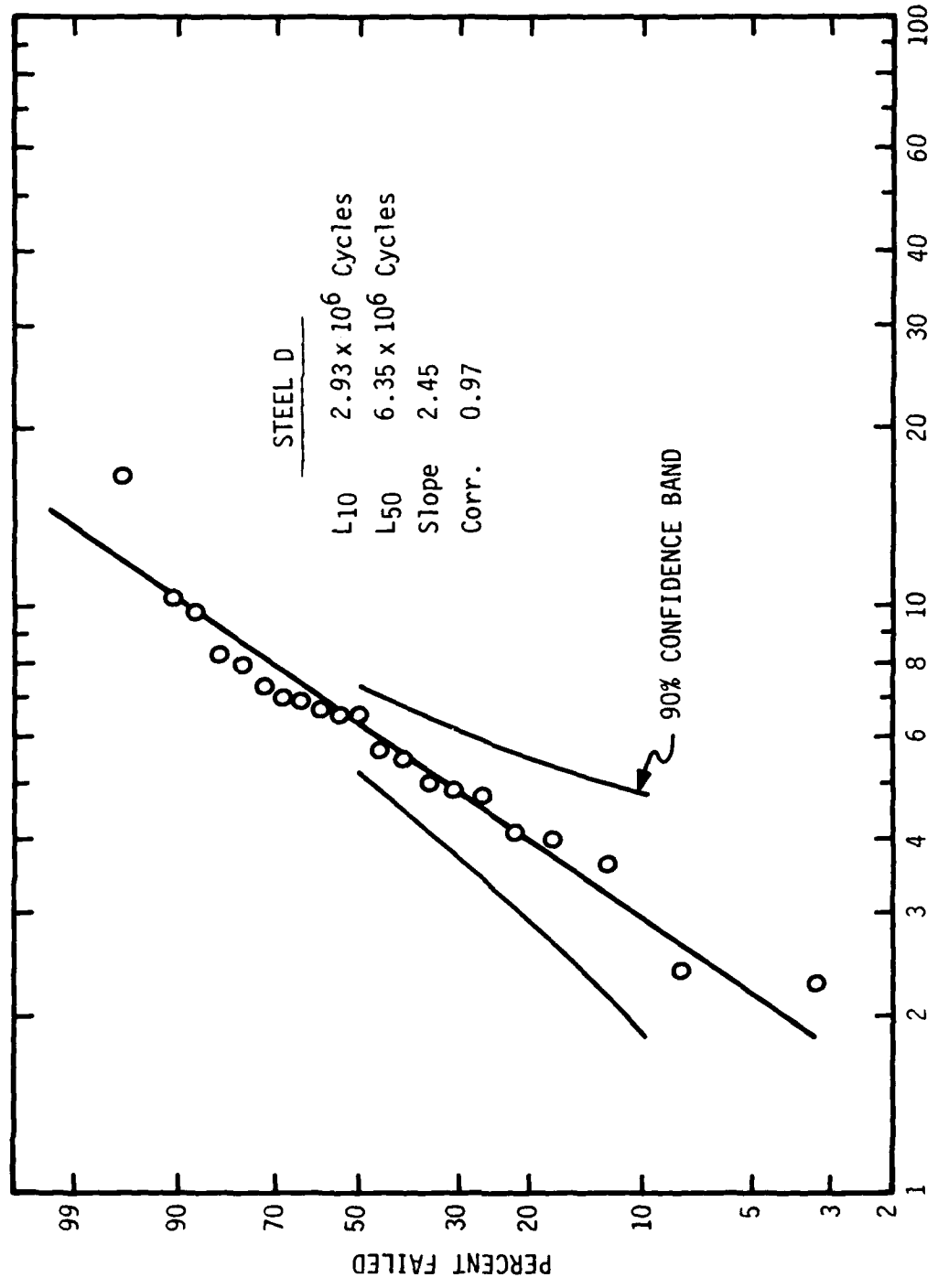


Figure 16 Rolling Contact Fatigue Results of Steel D

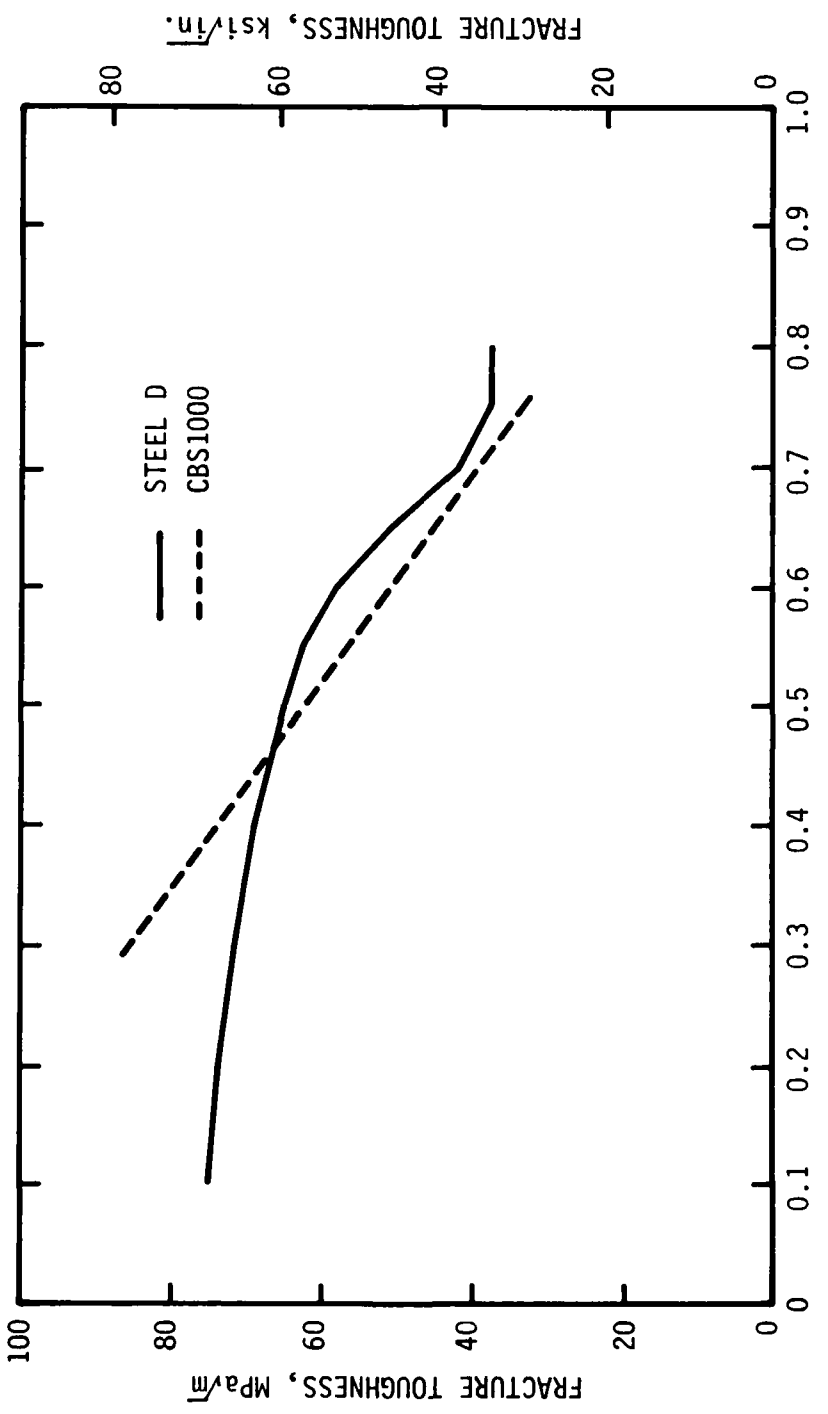


Figure 17 Comparison of Fracture Toughness Profiles for Steel D and CBS1000 (Ref. 1)

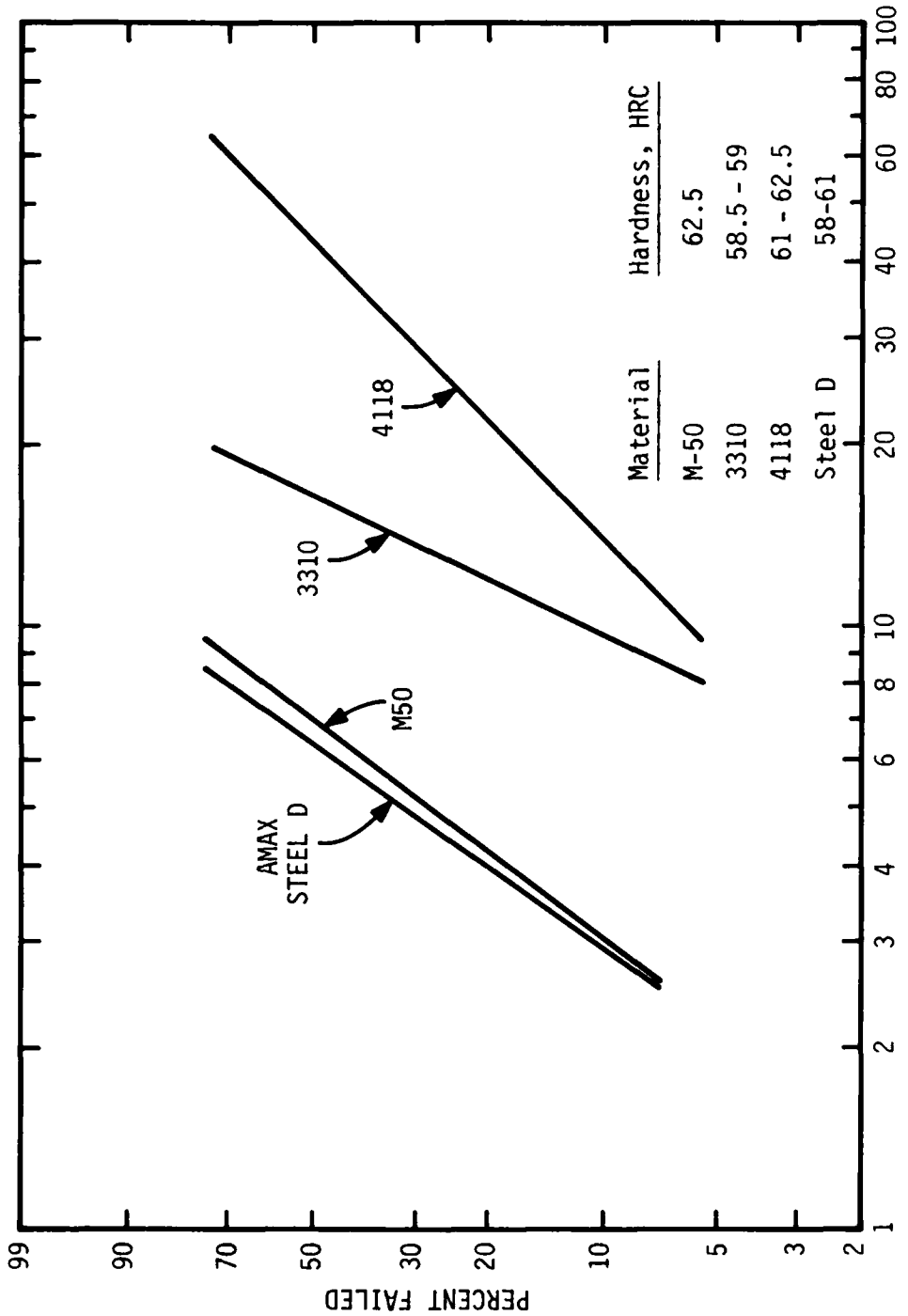


Figure 18 Rolling Contact Fatigue Results of Steel D Compared with Through Hardening M50 and Carburized SAE 3310 and 4118

APPENDIX

Fracture Toughness Comparison Between Steel D and CBS1000 in Figure 18

In the present investigation, the sides of the Charpy type fracture toughness specimens were "stopped off" to prevent carburization on the two opposite side faces. In the prior investigation,[1] the Charpy specimen had been carburized on all sides. The two processes result in a difference in the width of the analyzed volume at the base of the notched and precracked region. Whereas the analyzed volume in the present investigation has a width equal to the width of the specimen because carburization on side faces was prevented, the CBS1000 specimens have regions of higher carbon content (lower fracture toughness) at the extreme ends of the notched-precracked portion which, it is assumed, do not contribute to the measured fracture toughness of the lower carbon region. Hence, in order to compare the fracture toughness profile of the CBS1000 steel with Steel D, the fracture toughness values for CBS1000 (from Ref. 1) were adjusted as shown below:

$$K_{IC(adj)}^{CBS1000} = K_{IC} \times \frac{\text{Specimen Width}}{\text{Specimen Width} - 2\ (\text{Notch Depth})}$$

Because the assumption is made that the higher carbon regions at the extreme ends of the CBS1000 notched length made no contribution, the adjusted CBS1000 fracture toughness profile is probably somewhat on the high side of the true fracture toughness profile with the greater inaccuracy occurring in the lower carbon portions of the profile. However, this process does result in a conservative comparison of the fracture toughness differences between CBS1000 and Steel D, suggesting that the differences are probably somewhat larger than illustrated.

END

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