Factors Influencing Fracture Toughness of High-Carbon Martensitic Steels

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Abstract:

Plane strain fracture toughness of twelve high-carbon steels has been evaluated to study the influence of alloying elements, carbon content and retained austenite. The steels were especially designed to simulate the carburized case microstructure of commonly used automotive type gear steels. Results show that a small variation in carbon can influence the K_{IC} significantly. The beneficial effect of retained austenite depends both on its amount and distribution. The alloy effect, particularly nickel, becomes significant only after the alloy content exceeds a minimum amount. Small amounts of boron also appear beneficial.

Introduction

The issue of toughness of materials used for machine elements remains controversial, although much is understood about the subject. In the past few decades, our quantitative understanding of the subject has increased significantly. Concerned by brittle fractures in World War II liberty ships and, more recently, by rocket motor case failures, engineers have developed a comprehensive understanding of the concepts and test methods of fracture toughness and its measurement. Yet many myths concerning the use of toughness concepts in design persist. It is certainly an acceptable principle that designs must be strong and tough; however, application of strength and toughness concepts in the real world must be done through compromise, especially when fatigue is one of the design criteria.

Increased strength is usually accompanied by decreased toughness. In designs for which steel strengths below approximately 1700 MPa (40 R_C) are acceptable, material systems are available which offer appreciable toughness. However, in carburized applications, such as gears which are case-hardened to obtain a high-hardness, high-carbon martensitic case with a relatively low-hardness, low-carbon core, toughness becomes a specialized area of knowledge. Considerable controversy surrounds the problem of defining and evaluating the toughness required for heavy duty automotive gears.[1]

The present study was undertaken to obtain data useful

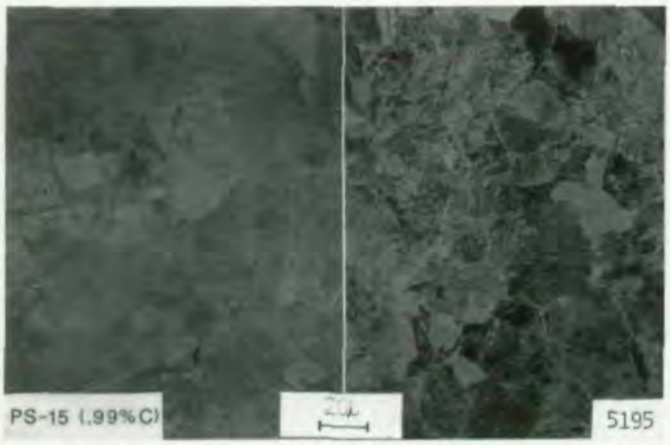

Fig. 1 – As-rolled microstructure in 0.99% C PS-15 and 5195 steels. Similar grain boundary carbide network surrounding pearlitic grains was developed in 8697, IH-50 and ERCH-1.

Table 1 – Chemical Composition

	C	MN	Cr	Ni	Mo	Si	Al	S	P
PS-15	0.99	1.09	0.54	0.03	0.16	0.32	0.05	0.02	0.02
PS-15	0.86	1.19	0.54	0.03	0.16	0.36	0.04	0.02	0.02
PS-15	0.72	1.11	0.51	0.03	0.17	0.28	0.04	0.02	0.02
8697	0.97	0.83	0.52	0.60	0.22	0.31	0.04	0.02	0.01
IH-50	0.97	1.37	0.05	0.03	0.01	0.68	0.05	0.02	0.02
4895	0.95	0.73	0.06	3.44	0.26	0.32	0.04	0.02	0.01
4870	0.70	0.76	0.05	3.50	0.25	0.31	0.04	0.02	0.01
ERCH-1	1.00	1.57	1.25	0.03	0.01	0.70	0.05	0.02	0.02
ER-8	0.95	1.32	0.95	1.57	0.01	0.31	0.03	0.02	0.01
9399	0.99	0.65	1.16	3.22	0.13	0.34	0.04	0.02	0.01
5195	0.95	1.00	0.93	0.03	0.01	0.31	0.04	0.02	0.01
PS15+B*	1.00	1.11	0.56	0.03	0.16	0.35	0.04	0.03	0.02

*PS15+B had 0.0008% B.
B content of all other steel < 0.0005%

in designing and selecting carburizing steels. The effects of alloying elements, carbon content and retained austenite on plane strain fracture toughness properties of high-carbon martensitic steels especially designed to simulate carburized case microstructure of several commonly used automotive steels were examined.

Procedure

Twelve 60 kg, high-carbon steel ingots were poured from air induction melted heats made with fine grain practice using aluminum and silicon. Chemical composition of the steels investigated is given in Table 1. All steels have a high carbon content to simulate the surface of a carburized component. The steels were designed to study the effect of alloying elements, case carbon content and retained austenite. To study the effect of carbon content on K_{IC}, Mn-Cr-Mo PS-15 steel was poured at 0.99%, 0.86% and 0.72% carbon, and Ni-Mo 4800 steel was produced at a carbon content of 0.99% and 0.70%. At equivalent carbon levels, SAE PS-15 steel is a cost-effective replacement steel for Mn-Cr-Ni-Mo 8600 series steel. Mn-Cr-Mo type steels have been used successfully in axles and power train components for over 15 years.

Comparison of the fracture toughness of 0.99% C PS-15 steel (residual nickel) with 8697 (intermediate nickel levels) provides information about the influence of low Ni on K_{IC} of high-carbon steels. IH-50 is a high-manganese, high-silicon steel. This type of steel is generally not used for carburizing applications because it has a relatively low case hardenability. The 4800 series steels, ERCH-1 and ER-8, were included to study the effect of nickel on fracture toughness. ERCH-1 and ER-8 are nickel-free and reduced nickel replacement steels (same case and base hardenability) for 4800 steels. The 4800 steel was poured at carbon contents of 0.95% and 0.70% to determine the effect of carbon content on K_{IC} of high-nickel, high-carbon steels. Other steels include 9300 steel with 0.99% C (9399), 5100 steel with 0.95% C (5195) and PS-15 with boron. The 9399 is a 1.16 Cr-3.22 Ni steel. The 5195 is a Mn-Cr steel with hardenability equivalent to high-carbon PS-15 and 8697 steels. Boron additions in Mn-Cr-Mo (PS-15 with boron) steel were made using commercially available Batts alloy.

The ingots were furnace-cooled and appropriate sections from each ingot were rolled into two 15 cm wide, 1 cm thick plates. Proper care was exercised to prevent cracking and distortion during the processing. Four to six compact tension specimens with width (W) equal to four times thickness (B) were machined from each grade of steel per ASTM E-399. The specimens were prepared in the "L-T" crack plane orien-

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tation. Before final machining, 3.75 x 3.75 x 1 cm pieces from the plates were ground to 0.6 cm. A minimum of 0.1 cm material was ground from each surface to remove any decarburized layer. Although all the data in this report were obtained using Standard ASTM E-399 compact tension specimens, tapered double cantilever beam (DCB) and oversized compact tension specimens (W=8B) were also used in the preliminary studies. DCB and oversized CT specimens resulted in 10-20% higher K_{IC} values. (See Table 2.)

Completely machined specimens were austenitized for two hours at 930°C, cooled to 845°C and vertically submerged into an oil quench tank at room temperature and tempered at 190°C for one hour. The surfaces of the specimens were protected from decarburization with a protective "decarb" stop coating. Representative heat treated specimens were checked for residual stresses, distortion and quench cracks. Quench cracks were not detected, and all specimens were within the dimensional tolerances permitted by ASTM E-399. Average residual stresses measured on the surfaces of two tapered DCB specimens were found to be +1.3 MPa. Considering the size and geometry of the compact tension specimen, it is assumed that the K_{IC} values reported herein are independent of macroresidual stresses.

The specimens were fatigue precracked, loaded to fracture in tension, and fracture toughness values were calculated from the autographic plot of load versus displacement in accordance with standard ASTM procedure. The K_{IC} values reported herein are averages of at least three replicate tests.

Since the specimens were heavily textured, the amount of retained austenite could not be determined by routine x-ray analysis. The volume fraction of retained austenite was measured by x-ray diffraction using a tilting and rotating stage.[2] The use of a rotating and tilting stage averages the x-ray peak intensity for a set of planes from a maximum number of possible crystallographic orientations.

Fractographic features on several selected compact tension specimens were studied using a scanning electron microscope. In addition, several compact tension specimens were sectioned and evaluated metallographically.

Results and Discussion

As-Rolled Microstructure.

The cleanliness level of all steels met typical JKT cleanliness requirements for commercial quality steels. The sulphides were pancake-shaped, having no abnormal size or distribution. Although each steel experienced a similar heating, rolling and cooling cycle, as-rolled microstructure of the plates varied significantly depending on the chemistry. (See Table 3.) Lower carbon, 0.72% C and 0.86% C, PS-15 steels developed a fully pearlitic microstructure. A proeutectoid grain boundary carbide network formed in 0.99% C PS-15, 8697, IH-50, ERCH-1 and 5195 steels. Typical as-rolled microstructures for 0.99% C PS-15 and 5195 steels are shown in Fig. 1. The as-rolled grain size in 0.99% C PS-15 steel was significantly coarser (ASTM grain size 1-2 as compared to 4-5) than all other steels showing proeutectoid carbides. The ER-8 steel had a primarily spheroid microstructure with small amounts of grain boundary carbides. High carbon 4895 steel developed a bainitic microstructure with patches of martensite. (See Fig. 2.) Fig. 2 also shows as-rolled microstructure for 9399 steel. Because of its high hardenability, 9399 steel developed a completely martensitic matrix with proeutectoid carbides in the grain boundaries.

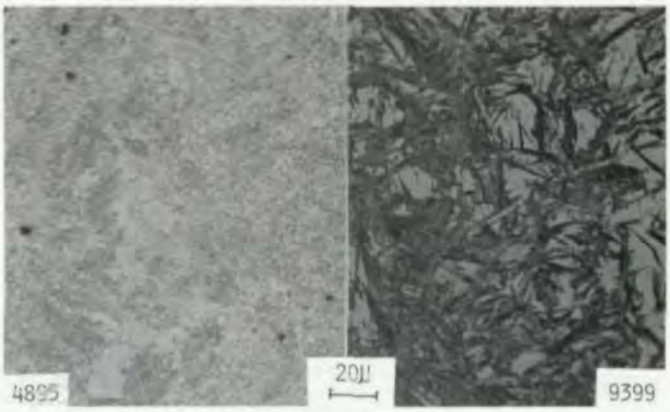

Fig. 2—As-rolled microstructures of 4895 and 9399 steels. Photomicrograph for 4895 steels shows essentially bainitic microstructure with patches of martensite. The 9399 steel developed proeutectoid carbides in a completely martensitic matrix.

As-rolled microstructure can influence subsequent heat treated mechanical properties of high-carbon, martensitic steels.[3],[4] Nakazawa and Krauss[4] evaluated the effect of proeutectoid carbides on the fracture toughness of quenched and tempered 52100 steel. Two series of specimens—one with grain boundary proeutectoid carbides in pearlitic matrix (Series A), and the other with well-distributed, fine spherical

Table 2 — Specimen Geometry Effect (High Carbon IH-50 Steel)

Hardness Rc	Fracture Toughness MPa.m$^{1/2}$ ASTM STD CT W=4B	Compact Tension W=8B	Tapered DCB
59.5	17.6	21.0	24.2
60.0	18.0	17.9	23.3
60.0	17.5	18.0	23.3
59.5	17.9	21.7	—
Average	**17.7**	**19.6**	**23.2**

Table 3 — As-Rolled Microstructures

Steel	Microstructure
PS-15 (.72% C, .86% C)	Fully Pearlitic
PS-15 (0.99% C)	Pearlitic With GB Carbides ASTM 1-2 GS
8697, IH-50, ERCH-1, 5195	Pearlitic With GB Carbides ASTM 4-5 GS
ER-8	Spheroidized With Small Amounts of GB Carbides
4895	Bainitic With Patches Of Martensite
9399	Martensitic With Proeutectoid Carbides

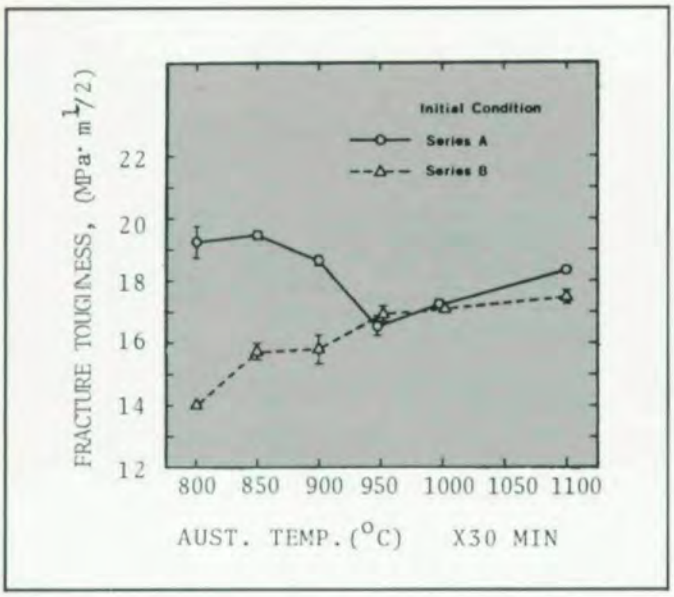

Fig. 3 – Influence of starting microstructure on K_{IC} of 52100 steel. Series A had proeutectoid carbides. Series B contained only fine carbides.[4]

carbides in a spheroidized matrix (Series B) – were produced, and their fracture toughness was determined as a function of the austenitizing temperature. The results are reproduced in Fig. 3. When austenitized below 950°C, martensitic specimens with proeutectoid carbides in the starting microstructure (Series A) are significantly tougher than Series B, which contained only fine spherical carbides. Above

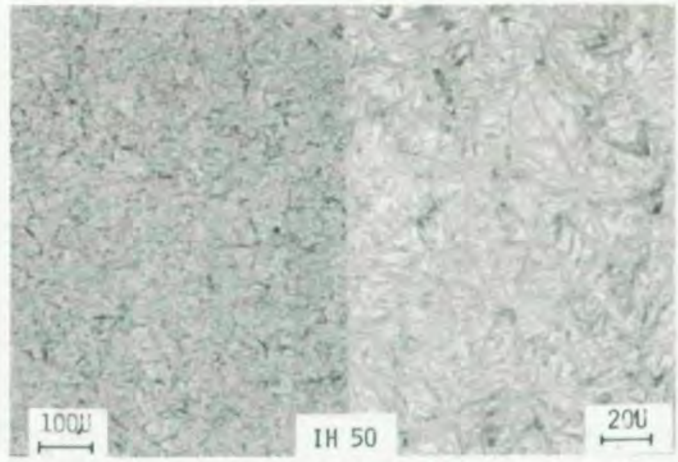

Fig. 4 – Microstructure of high carbon IH-50 steel. The microstructure shows a uniform, lightly tempered, high-carbon martensite and retained austenite with no carbides or microcracks.

950°C, where all carbides dissolve to form a homogenous austenite, both series have about the same toughness. Nakazawa and Krauss explained the increased toughness of Series A on the basis of higher toughness of carbide-free, transgranular areas present between the proeutectoid carbides. Below 950°C, specimens containing only fine spheroid particles have low toughness because of the very fine microvoid coalescence (small ligaments of tough carbide-free areas) associated with the closely spaced carbides. In the present studies, all specimens were austenitized at 930°C, where essentially all proeutectoid carbides were dissolved. Therefore, variations in as-rolled microstructure are expected to have little or no influence on the K_{IC} values.

Heat Treated Microstructure.

Typical microstructures for various steels after the heat treatment (austenitized at 930°C for two hours, quenched in oil from 845°C and tempered at 190°C for one hour) are shown in Figs. 4-7. As shown in Fig. 4, IH-50 developed a uniform microstructure containing lightly tempered, high-carbon martensite and retained austenite with no carbides or microcracks. This is the type of surface microstructure which is most desirable in carburized parts. However, as indicated previously, IH-50 (high-manganese, high-silicon) is not a suitable steel for most automotive gearing and shaft applications because it does not have sufficient case hardenability.

With the exception of IH-50 steel, all other steels exhibited non-uniform, banded microstructures. The severity of carbon and alloy segregation varied depending upon the type

Table 4 – Undissolved Carbides
QIA 720 Measurements

Steel	% Carbides
PS-15 (.99% C)	1.0%
8697	1.0%
ERCH-1	9.0%
ER-8	1.5%
9399	8.5%
5195	1.0%
PS-15+B	1.0%

.72% C PS-15, .86% C PS-15, 4870 and 4895 contained no undissolved carbides.

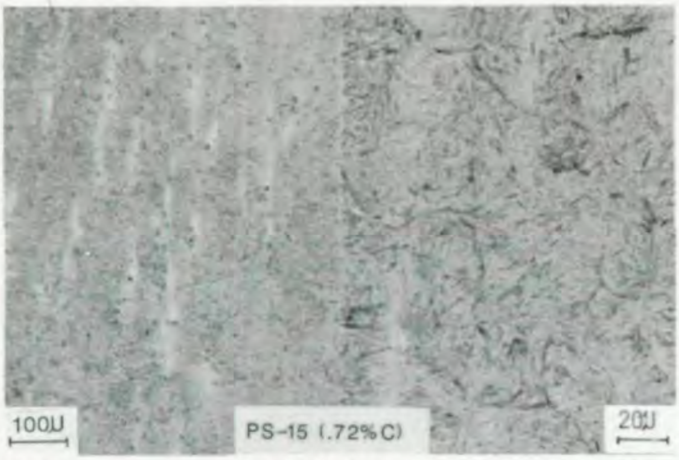

Fig. 5 – Microstructure of 0.72% C PS-15 steel. The white etching bands are higher carbon areas which are relatively soft and contain higher amounts of retained austenite. No carbides or microcracks are visible.

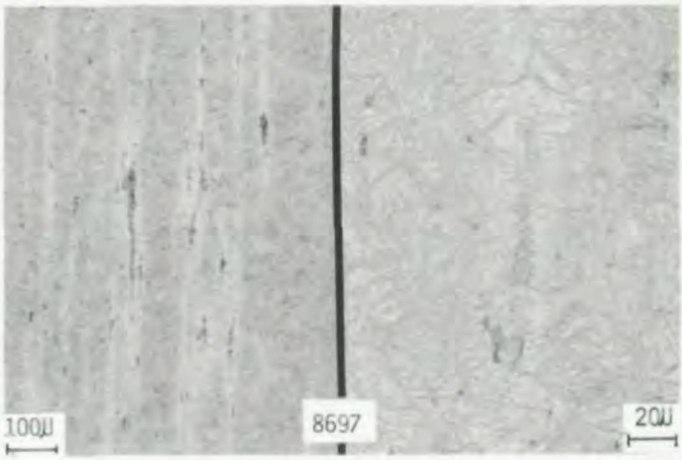

Fig. 6 – Microstructure of 8697 steel. PS-15 with 0.99% C also showed a similar microstructure. In addition to the microcracks, the white etching bands contained undissolved carbides.

and amount of alloying elements. PS-15 steels with 0.72% and 0.86% carbon, when examined at 100X, showed numerous white etching bands (Fig. 5). The white etching bands were relatively soft (HRC 62 vs 65) and contained a higher volume fraction of retained austenite as compared to the surrounding matrix. Carbides or microcracks were not observed in the 0.72% C PS-15 steel. The 0.86% C PS-15 steel also did not have carbides, but occasionally microcracks were observed. Microstructures of high-nickel 4870 and 4895 steels were similar to the 0.72% PS-15 steel. High-nickel steels also showed non-uniform microstructure with no carbides or microcracks. The frequency of white etching bands, however, was low. The martensitic plate size in the nickel steels was somewhat coarser.

The 8697 and 0.99% C PS-15 steels had similar microstructures. In addition to microcracks, the white etching bands contained numerous undissolved carbides. A typical microstructure for heat treated 8697 is shown in Fig. 6. The worst situation with respect to carbides and banding was observed in ERCH-1, a nickel-free substitute steel for the 4800 grade of steel (Fig. 7). The microstructure of 9399 steel was very similar to ERCH-1.

The amount of undissolved carbides was determined using a quantitative image analyzing system. As illustrated in Table 4, 0.99% C PS-15, 8697, 5195 and PS-15+B steels have less than 1% carbides. ER-8, a substitute steel for 4800 steel with reduced nickel content, had 1.5% undissolved carbides. The highest amount of carbides, 8.5 and 9%, were present in 9399 and ERCH-1 steel, respectively.

The influence of carbides on K_{IC} was investigated by determining the fracture toughness of ERCH-1 steel after austenitizing it at 930°C for two, four and seven hours. Samples of carbide free 0.86% C PS-15 steel were also included in the study for comparison. Increasing the austenitization time from two to seven hours did not change fracture toughness of either steel significantly (Table 5). The influence of massive carbides on fracture toughness of austenitic and martensitic white cast iron has been studied by Gahr and Scholz.[5] On the basis of the results shown in Fig. 8 from that study, the authors concluded that increasing the volume of carbides from 7% to 30% had no significant influence on K_{IC}. Only when the volume of carbide increases beyond 30%, did the fracture toughness start decreasing. The matrix of martensitic white cast iron is essentially the same as the high carbon steel. None of the steels included in the present investigation contained more than 9% undissolved carbides. In fact,

Table 5 – Effect of Austenitizing Time at 930°C

	K_{IC} MPa.m$^{1/2}$		
Steel	2 Hours	4 Hours	7 Hours
PS-15 (.87% C)	22.4	23.8	21.8
ERCH-1	20.7	21.1	20.0

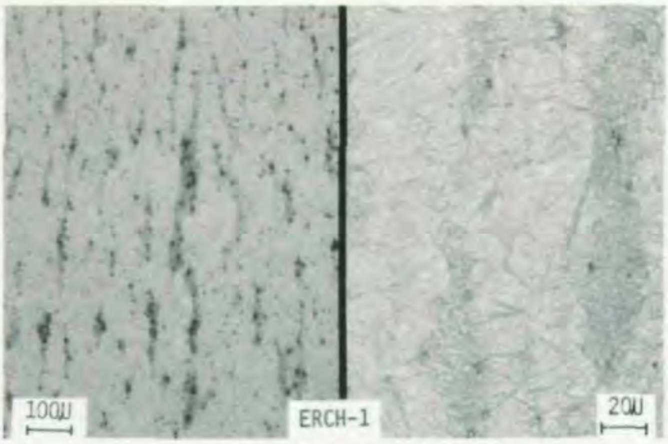

Fig. 7—Microstructure of ERCH-1 steel. ERCH-1 steel showed maximum banding with greatest amount of undissoved carbides.

as given in Table 4, the majority of steels have less than 2% undissolved carbides. It is, therefore, reasonable to assume that the fracture toughness values reported herein have not

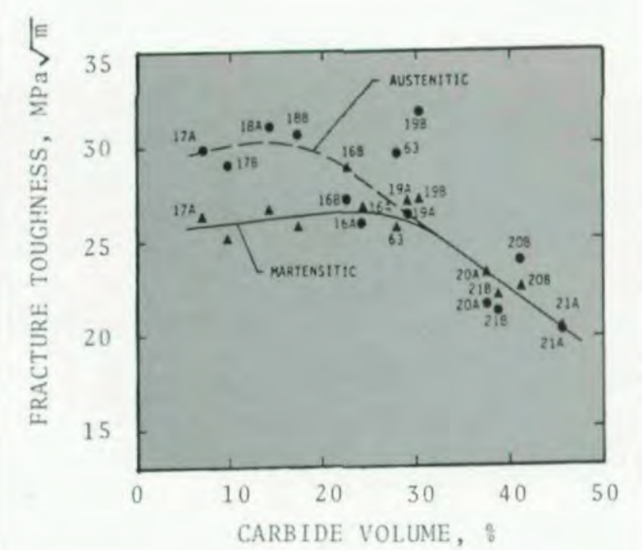

Fig. 8—Influence of carbides on fracture toughness of martensitic and austenitic white cast irons.[5]

been influenced by the presence of carbides. Since more than 90% of the fracture propagates through a martensite-austenitic matrix, fracture toughness values should be dominated by the matrix properties rather than the carbides.

Influence of Retained Austenite.

The presence of retained austenite, particularly as a continuous thin film surrounding martensitic plates rather than as a discreet blocky phase, is considered to enhance fracture toughness of martensitic steels. Thomas[6] has emphasized the importance of mechanical and thermal stability of retained austenite. Retained austenite has been proposed to improve crack propagation resistance and, thereby, increase fracture toughness by crack branching or blunting,[7] by strain or stress induced transformation resulting in the development of compressive stresses in front of the advancing crack,[8] by preventing the formation of brittle boundary carbides, and by breaking the continuity of the cleavage planes across various martensitic plates.[9]

The influence of retained austenite on fracture toughness of high carbon steels was studied by varying the amount of retained austenite in 4895 steel by deep freezing at $-84°C$ and $-207°C$. The results in Table 6 show that a decrease in the austenite from 40% to 18.5%, caused by deep freezing, quenched and tempered 4895 steel at $-84°C$, decreased fracture toughness from 24.5 MPa.m$^{1/2}$ to 14.6 MPa.m$^{1/2}$. Further deep freezing at $-207°C$ reduced retained austenite to 15.5% causing an additional decrease in the toughness to 12.5 MPa.m$^{1/2}$. The amount of retained austenite, therefore, has a significant effect on the K_{IC} of high-carbon, martensitic steels. The microstructure of 4895 steel after the subzero treatments is shown in Fig. 9. Deep freezing transforms retained austenite to martensite. A careful examination of the microstructure revealed that additional microcracks were generated as a result of the transformation. A 50% decrease in the K_{IC} value, however, cannot be explained on the basis of an increase in the density of microcracks. This decrease is primarily related to the reduction of retained austenite. Suf-

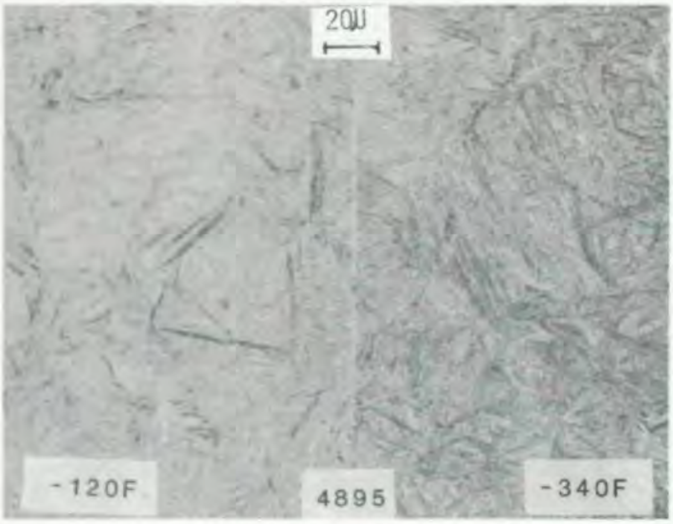

Fig. 9 – Microstructure of 4895 steel austenitized at 930° C, quenched in oil, tempered at 190° C for 1 hour and deep frozen at −84° C and −207° C for two hours.

ficient data were not generated to determine the minimum level of retained austenite necessary to increase toughness significantly. The beneficial effect of retained austenite depends on its nature and distribution. Retained austenite is most effective when present as continuous films surrounding martensite rather than a discontinuous blocky phase.[9] Heat treatment and alloying elements could be selected to maximize the influence of retained austenite at lowest levels necessary so that the toughness can be improved without detrimentally affecting the hardness and other related properties.

Effect of Carbon Content.

Increased strengthening caused by carbide precipitation or interstitial solid solution strengthening by carbon is usually accompanied by decreased toughness. In designs for which steel strength below 1700 MPa (40 HRC) are acceptable, material systems are available offering appreciable toughness. In high-hardness, carburized applications, (2000 MPa (57 HRC) and above) such as gears, toughness becomes a specialized area of knowledge. Earlier work in this area was done by Schwartzbart and Sheehan.[10] Results of some of their work, replotted, are shown in Fig. 10. This work was done with Charpy V-notch specimens. The same hardness at different carbon levels was obtained by tempering.

In the present investigation, the effect of carbon on K_{IC} of high-carbon, martensitic steel was studied by testing PS-15 steel at 0.99%, 0.86% and 0.72% carbon. Also, the fracture toughness of 4800 steel was determined at 0.95% and 0.70% carbon. The results are shown in Table 7. Lowering carbon enhances fracture toughness. The effect of carbon, however,

Table 6 — Effect of Retained Austenite 4895 Steel

Heat Treatment	Rc	RA	K_{IC}MPa.m$^{1/2}$
190°C Temper	55.5	40.0%	24.5
Deep Frozen -84°C	64.0	18.5%	14.6
Deep Frozen -207°C	65.0	15.5%	12.5

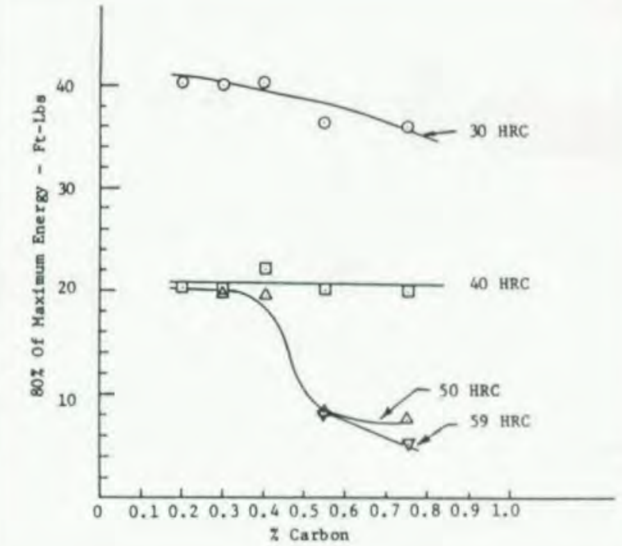

Fig. 10 – Isohardness relation of carbon and toughness. Toughness determined by Charpy V-notch.[10]

must be considered in conjunction with retained austenite. In general, the amount of retained austenite decreases with decreasing carbon. In the case of PS-15 steels, a change in the carbon from 0.99% to 0.86% decreased retained austenite from 39% to 23%. Discounting other effects, this

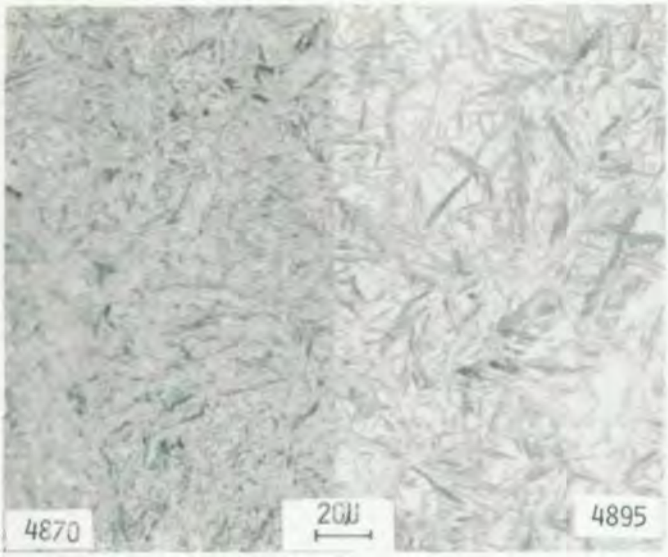

Fig. 11 – Microstructure of 4870 and 4895 steels.

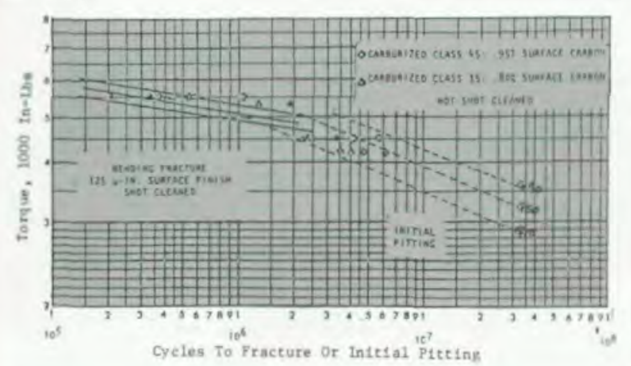

Fig. 12 – Comparison of bending and contact fatigue life durability of carburized Mn-Cr steel with G10, G50, and G90 (solid and dotted lines) bands for carburized 8622 steel.

reduction should have lowered the toughness. However, the toughness increased, indicating the importance of carbon. K_{IC} decreased slightly when the carbon content was reduced from 0.86% to 0.72% C. Improved toughness resulting from lowering the carbon in this case is more than offset by a decrease in the toughness caused by a change in the retained austenite from 23% to 16%.

The effect of carbon on fracture toughness of high-carbon steels becomes more evident in the case of high-nickel 4800 series steels. A decrease in the carbon from 0.95% to 0.70% increased K_{IC} from 24.5 MPa.m$^{1/2}$ to 34.5 MPa.m$^{1/2}$. Nickel promotes the stablization of retained austenite. Therefore, even at 0.70% C, 4870 steel containing 3.50% nickel retained 21% austenite. The highest fracture toughness was recorded

Table 7 – Effect of Carbon Content on K_{IC}

Steel	%C	Rc	RA	K_{IC}MPa.m$^{1/2}$
PS-15	0.99	60.0	39%	16.6
PS-15	0.86	60.5	23%	22.4
PS-15	0.72	60.5	16%	21.7
4895	0.95	55.5	40%	24.5
4870	0.70	57.0	21%	34.5

for this steel. Notice that the fracture toughness improved in spite of an increase in the hardness by nearly 2 Rc.

Lower carbon martensites are inherently more ductile. It is well known that the morphology of martensite changes from dislocated laths to twinned plates with increasing solute content. Carbon has the strongest effect in promoting twinning. Alloy composition and martensitic transformation temperature influence the transition of martensite from lath to plate martensite. The dislocated laths are composed of bundles of slightly misoriented grains having high density of tangled dislocations. The fine structure of plate martensite consists of fine transformation twins.[11] Lower carbon dislocated lath martensites are relatively less brittle, because under stress, dislocations can move, resulting in deformation by slip. Higher-carbon martensite, which consists primarily of plate martensite with internal twins, is brittle. Twinned martensites have extremely low fracture toughness. A reduction in the percentage of twinned martensite, resulting from lowering carbon content, can be expected to improve K_{IC} significantly. It has been shown that plane strain fracture toughness of AISI 4130 (medium-carbon low-hardenability steel) can be improved two-fold by a high temperature and step-cooling austenitizing treatment, which virtually eliminates twinned martensite plates.[12]

The microstructures of 4895 and 4870 steels are shown in Fig. 11. Photomicrographs show a difference in the martensitic plate size and the amount of retained austenite. The 4870 has finer plate size and less retained austenite. It is also likely that the 4870 steel contains lower amounts of plate martensite, although the structure for both steels is mostly plate type.

Alloy Effects.

Our past experience indicates that in automotive gear steels, small amounts of nickel and molybdenum can be replaced with other alloying elements. A basic assumption in the development of cost-effective carburizing steel is that engineering performance of a component made from low alloy steel is dictated by the base and case carbon content, microstructure and residual stresses. As long as the steel selected meets these requirements, equivalent engineering performance can be expected. The microstructure and residual stresses are governed by the carbon content, hardenability and martensite start and finish transformation temperatures. The major reason for alloying these steels is to influence hardenability and transformation temperatures. Sufficient hardenability, coupled with adequate and uniform quenching, provides microstructure optimization and control. The dependence of performance/strength on microstructure is basic.

Considerable laboratory test and field history data exist to support these basic assumptions. SAE PS-16 (Mn-Cr-Mo) steel, which is a nickel-free, cost-effective replacement steel for standard AISI 8622 (Mn-Cr-Mo-Ni) steel, has been used successfully in the production of heavy duty truck rear axle ring gears and in tractor and combine power trains for over 15 years. SAE PS-59, a cost-effective, Mn-Cr replacement steel, has also been used to substitute for PS-16 and 8622 steel for the last five years. Fig. 12 shows torque-life gear data from PS-59 steel at two carbon levels, superimposed on the G10, G50 and G90 bands for 8600 type steel. The bending and contact fatigue durability data were obtained by testing carburized and hardened six-pitch test pinions in a power circulating rig. The results show that the Mn-Cr steel and Mn-Cr-Mo-Ni (8622) steel have equivalent fatigue properties.

The data in Fig. 12 indicates that in low alloy gearing steels, small amounts of nickel and molybdenum can be replaced with other more cost-effective alloying elements. Nonetheless, the effect of alloying elements on the performance of gears is quite controversial. Using a modified Brugger test, Diesburg[13] studied the influence of carbon, alloy and residual stress on fracture stress and very low cycle fatigue life. Carbon and residual stress were shown to be very important. Alloy content was also found to be important, but the amount of alloy appeared more important than the specific alloy. DePaul,[14] Shea[15] and Love and Campbell,[16] among others, provide a technology base concerning the effect of major alloys on properties of carburized steel. Most of this relatively small amount of work on alloy effects has been related only to gear tooth bending fatigue strength, not gear tooth flank durability (pitting fatigue). Krauss[17] concluded, based on roller specimens, that the alloying elements alone had no effect on flank durability. Allsopp, Weare and Love [18] also were not able to detect an alloy influence in tests using gears. None of the above programs, however, were designed specifically to systematically study alloy effects.

Effect of Nickel at Intermediate Levels.

Fracture toughness K_{IC} for 0.99% C PS-15 steel, Mn-Cr-Ni-Mo steel 8697 and 5195 steels are given in Table 8. At equivalent carbon levels, K_{IC} for PS-15 and 8600 series steels are exactly the same; i.e., 16.6 $MPa.m^{1/2}$. The 8695 steel has a slightly lower retained austenite. However, the amount of austenite is sufficient to provide its maximum contribution to toughness. Fracture toughness of 5195 Cr-Mn steel is superior to both the high carbon PS-15 and 8699 steels. This improved toughness is thought to be related to a lower matrix carbon concentration in 5195 steel caused by the formation of chromium carbides. The data in Table 8 show that small amounts of nickel, molybdenum and other alloying elements do not influence the fracture toughness of high carbon steels significantly.

Table 8 – Fracture Toughness of High Carbon PS-15, 8600 and 5100 Steels

Steel	C	Mn	Cr	Ni	Mo	HRC	% RA	K_{IC} $MPa.m^{1/2}$
PS-15	0.99	1.09	0.54	0.03	0.16	60	39	16.6
8697	0.97	0.83	0.52	0.60	0.22	60	33	16.6
5195	0.95	1.00	0.93	0.03	0.01	60	34	20.8

Effect of Nickel at Higher Levels.

Table 9 shows the effect of nickel on fracture toughness of high-carbon martensitic steels. A comparison of K_{IC} for 4895, ER-8 and ERCH-1 reveals that fracture toughness increases with increasing nickel content. ER-8 and ERCH-1 are reduced nickel and nickel-free replacement steels for 4800 steel. K_{IC} of ER-8, in which half of the nickel in 4800 steel is replaced with other alloying elements, has approximately 10% lower fracture toughness as compared to the 4895 steel. ERCH-1, which contains no nickel, has the lowest toughness. The effect of nickel can also be seen by comparing the K_{IC} of 0.72% C PS-15 steel with 4870 and 9399 steel with 5195. Although some of the improvement in toughness may be explained on the basis of higher amounts of retained austenite and lower hardness of 4870 and 5195 steels, a nominal 3.5% nickel in these steels seems to promote improved toughness.

Effect of Boron.

Limited data obtained on testing a steel containing boron reveals that small additions of boron can enhance fracture toughness of high-carbon martensitic steels significantly. (See Table 10.) The addition of 0.0008% boron to 0.99% C PS-15 steel improved fracture toughness from 16.6 $MPa.m^{1/2}$ to 21.2 $MPa.m^{1/2}$. Boron is known to be a potent contributor to hardenability in the low and medium carbon range, but has very little influence on the hardenability of high-carbon steels. This non-hardenability related improvement in the

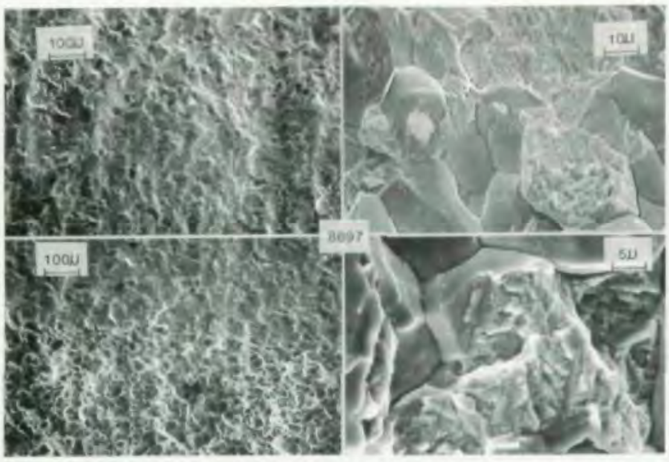

Fig. 13 – Photomicrographs showing fracture modes in fatigue precrack and single overload regions of 8697 steel. (A) Fatigue precrack area, (B) and (C) fatigue crack/single overload interface, and (D) transgranular area in single overload.

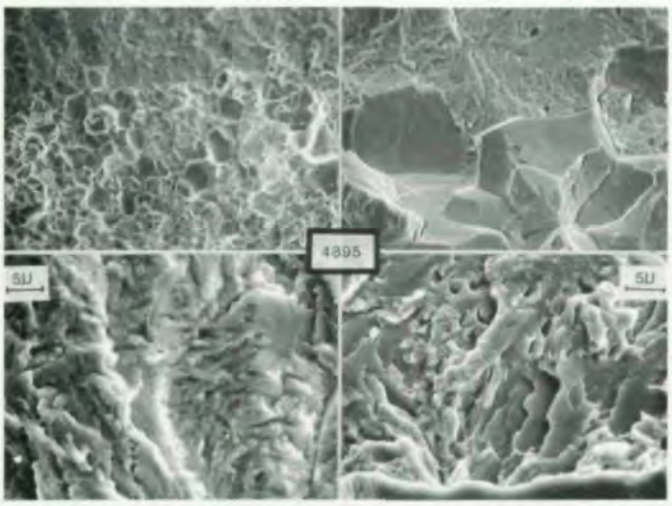

Fig. 14 – Photomicrographs showing fracture modes, fatigue precrack and single overload regions of 4895 steel. (A) and (C) fatigue precrack/single overload interface, (B) transgranular area in fatigue precrack area, and (D) transgranular area in the overload region.

Table 9 — Effect of Nickel Content On K_{IC}

Steel	%C	% Ni	Rc	RA	K_{IC}MPa.m$^{1/2}$
4895	.95	3.44	55.5	40%	24.5
ER-8	.95	1.57	60.5	47%	22.4
ERCH-1	1.00	0.08	58.5	42%	20.7
PS-15	.72	0.03	60.5	16%	21.7
4870	.78	3.50	57.0	21%	34.5
9399	.99	3.22	53.0	49%	27.1
5195	.95	0.08	60.0	34%	20.8

Table 10 — Effect of Boron On K_{IC}

Steel	%C	Rc	RA	K_{IC}MPa.m$^{1/2}$
PS-15	0.99	60	39%	16.6
PS-15 + B*	1.00	60	31%	21.2

*Added as Fe-Al-B (BATTS #2) Alloy.

Table 11 — Critical Crack Size and Load Carrying Capability of Various Steels

STEEL (%C)	K_{IC}(MPa.m$^{1/2}$)	Ac (mm)	σc (MPa)(1)	R(2)
PS-15 (.99)	16.6	0.363	690	1.00
PS-15 (.86)	22.4	0.660	930	1.35
PS-15 (.72)	21.7	0.615	900	1.30
8697 (.97)	16.6	0.363	690	1.00
IH-50 (.97)	20.3	0.543	840	1.22
4895 (.95)	24.5	0.787	1015	1.47
4870 (.70)	34.5	1.520	1410	2.04
ERCH-1 (1.00)	20.7	0.566	860	1.25
ER-8 (.95)	22.4	0.660	930	1.35
9399 (.99)	27.1	0.965	1125	1.63
5195 (.95)	20.8	0.571	870	1.26
PS-15 + B (1.00)	21.2	0.589	885	1.28

(1) Critical applied stress for a crack size of 0.363 mm.
(2) R = $\sigma c/\sigma$ for PS-15.

fracture toughness of high-carbon boron steel may depend on how boron additions are made. Unlike the domestic practice of alloying with "protected" boron by simultaneously adding titanium, zirconium and aluminum, the German practice is to alloy with ferroboron. The German practice is an inefficient way to use boron for hardenability enhancement, but the practice is said to improve the distortion, toughness and low and high cycle fatigue characteristics of carburized steel.[19] According to the German practice, when added as a ferroboron, boron acts as a nitrogen fixer during both the steel making process and subsequent carburizing treatment. After combining with nitrogen during the steel making process, enough boron remains in the soluble form to combine with nitrogen picked up in carburizing and, thus, high carbon martensite-austenite surface is low in soluble nitrogen and has enhanced properties.[20] Additional research is required to fully assess non-hardenability related beneficial effects of boron-containing carburized steels.

Significance of Fracture Toughness in Design.

Small differences in the K_{IC} resulting from carbon content, retained austenite or alloy effects can have a significant influence on the critical size (A_c) and load carrying capability (σ_c) of high-carbon steels. The data in Table 11 were calculated using basic fracture mechanics concepts involving relationships between stress intensity factor, applied stress and crack length. The critical size (A_c) was calculated for an elliptical surface crack with a geometry A/2c of 0.5, assuming applied stress of 690 MPa and applied stress-to-yield strength ratio of 0.33. The results show that a change in the K_{IC} of high-carbon steel from 16.6 to 22.4 MPa.m$^{1/2}$ nearly doubles the size of the crack that the material can sustain before an unstable crack growth occurs. Similarly, for a given crack size, the load carrying ability of the steel improves by 35%.

Fractography.

The fatigue precrack and single overload regions of several samples were examined using SEM. Typical fractographs are shown in Figs. 13-15. Stable fatigue crack growth in the precrack area was characterized by both transgranular and intergranular fracture modes. In all samples, fracture topography at low stress-intensity range was smooth and primarily transgranular. As the stress-intensity range in-

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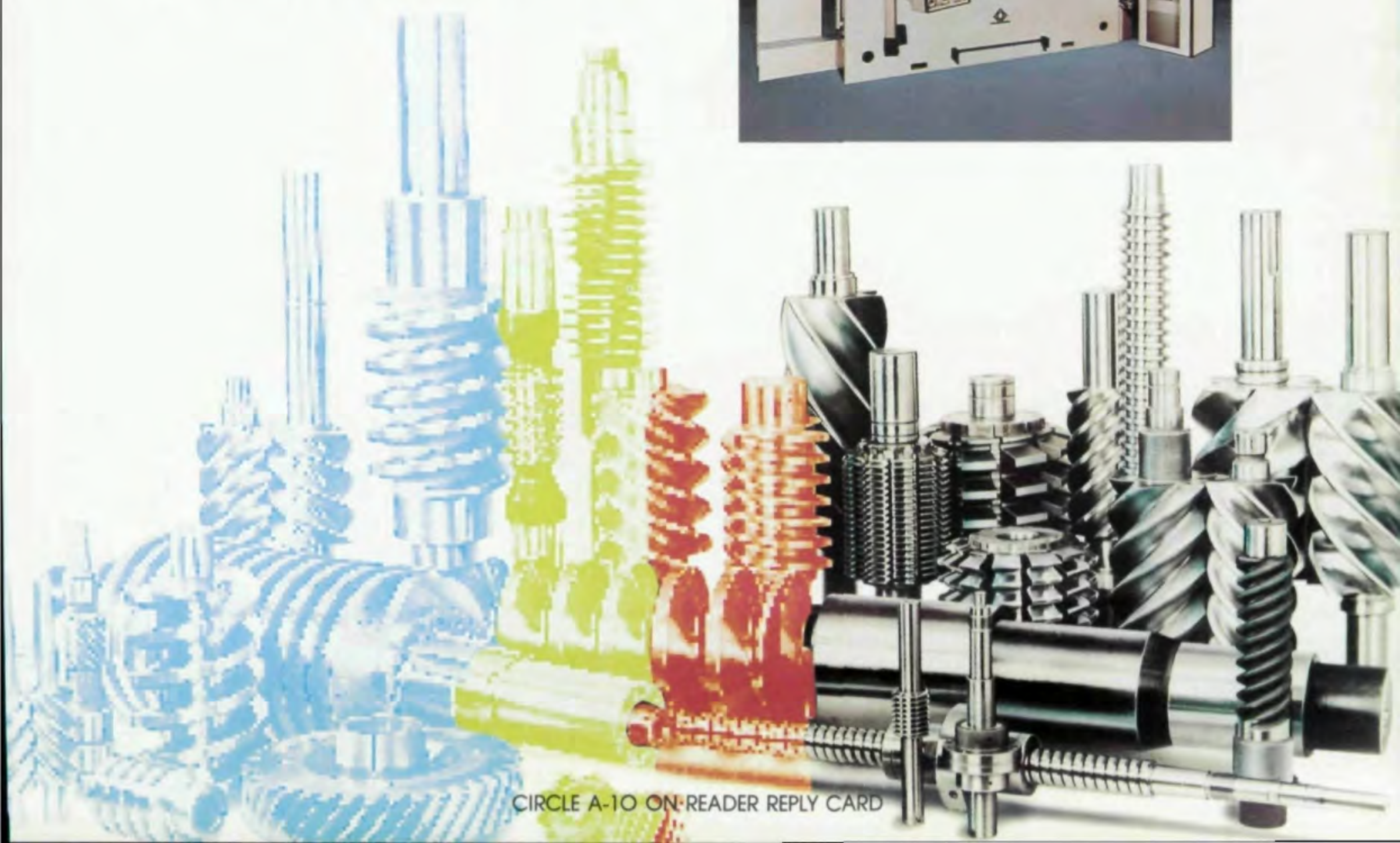

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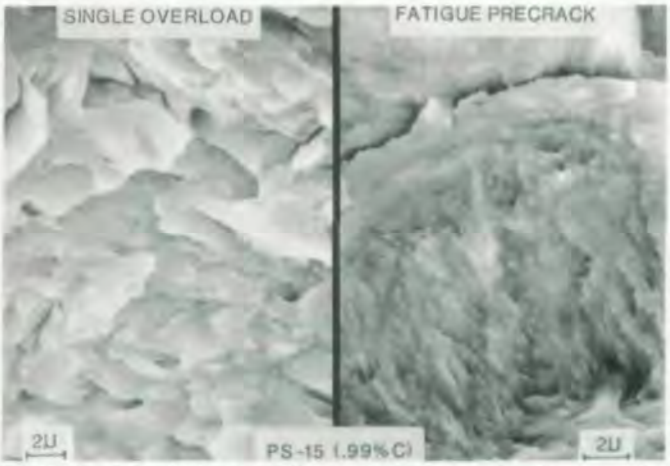

Fig. 15 – Fractographs showing differences in the topography of transgranular fractures in the single overload and fatigue precrack regions.

creased, the fracture surface generally became coarser, and the percentage of transgranular areas decreased, while the amount of intergranular fracture mode increased. Further examination of the transgranular area at higher magnification revealed topography typically associated with crack propagation in high-hardness steels. The fracture surfaces were flat and mottled, but occasionally fatigue striation could be resolved.

The single overload region was primarily intergranular with some transgranular areas present. These transgranular areas were distinctively different from the topography observed in the transgranular fatigue precrack areas. (See Fig. 15.) These transgranular areas were characterized by cleavage and flat dimples.

Conclusion

The results show that a complex interdependency existing between the carbon content, retained austenite and various alloying elements makes interpretation of the data difficult. Experimental steels must be designed carefully to isolate the effect of each of these variables.

A small variation in the carbon can influence the fracture toughness of high-carbon steels significantly. Lowering carbon increases toughness. The effect of carbon, however, must be considered in conjunction with retained austenite, which increases with the increasing carbon content. Higher retained austenite is usually beneficial, but the beneficial effect is dependent on its nature and distribution.

The alloy effect becomes significant only after the alloy exceeds a minimum amount. Results show that small amounts of nickel ($\sim 0.5\%$), molybdenum ($\sim 0.2\%$) and other alloying elements do not have significant effect on the plane strain fracture toughness properties of high-carbon steels. Higher nickel promotes toughness. K_{IC} of high-carbon steels containing a nominal 1.5% and 3.5% nickel improved significantly. Small amounts (0.0008%) of boron also appear beneficial. This non-hardenability related improvement may depend on how boron additions are made.

References:

1. RAZIM, C. and STRENG, H. "Properties of Case-Hardened Components – Basic Considerations Concerning the Definition and Evaluation of Toughness," *Proceedings, Case-Hardened Steels: Microstructural and Residual Stress Effects*. D.E. Diesburgh, Ed. AIME, 1983, p. 1-15.
2. SHIN, S.W. and SHARMA, V. K. "Application of a Tilting and Rotating Specimen Stage to X-Ray Retained Austenite Measurements in Textured and Coarse Grained Steels," SAE Paper #800428, *SAE Transactions*, 1980.
3. STICKELS, C.A. "Plastic Deformation of Quenched and Tempered 52100 Steels Under Compression," *Metallurgical Transactions A*, 1977, Vol. 8A, pp. 63-70.
4. NAKAZAWA, K. and KRAUSS, G., "Microstructure and Fracture of 52100 Steel," *Metallurgical Transactions A*, Vol. 9A, 1978, pp. 681-689.
5. GAHR ZUM, K.H., and SCHOLZ, W.G. "Fracture Toughness of White Cast Irons," *Journal of Metals*, October, 1980, p. 38.
6. THOMAS, G. "Retained Austenite and Tempered Martensite Embrittlement," *Metallurgical Transactions A*, Vol. 9A, March, 1978, p. 439.
7. WEBSTER, D."Development of a High Strength Stainless Steel with Improved Toughness and Ductility," *Metallurgical Transactions* Vol. 2, 1971, p. 2097.
8. GERBERICH, W.W., et al. *Fracture 1969*, Proc. of 2nd. International Conference on Fracture, Brighton, U.K. Chapman and Hall, Ltd., 1969, p. 288.
9. KOO, J., RAO, B.V.N., and THOMAS, G. "Designing High Performanace Steels with Dual Phase Structures," *Metal Progress*, September, 1979, p. 67.
10. SCHWARTZBART, H. and SHEEHAN, J. "Impact Properties of Quenched and Tempered Alloy Steel," *Final Report ONR Contract N60NR244T.O.I.*, September, 1955, Research conducted at IIT, Chicago.
11. KRAUSS, G. and MARDER, A.R. "The Morphology of Martensite in Iron Alloys," *Metallurgical Transactions*, Vol. 2, September, 1971, p. 2343.
12. ZACKAY, V. F., "Fundamental Consideration in the Design of Ferrous Alloys," *Alloy Design for Fatigue and Fracture Resistance, AGARD-CP-185*. NATO, January, 1976.
13. DIESBURG, D.E. and SMITH, Y.E. "Fracture Resistance in Carburized Steels, PE II: Impact Fracture," *Metal Progress*, 115 (6), June, 1979, pp. 35-39.
14. DEPAUL, R.A. Impact Fatigue Resistance of Commonly Used Gear Steels, SAE paper 710277, 1971.
15. SHEA, M.M. Impact Properties of Selected Gear Steels, SAE Paper 780772, 1978.
16. LOVE, R.A. and CAMPBELL, J.G. "Bending Strength of Gear Teeth," Report 1925/5, Motor Industry Research Association of U.K., December, 1952.
17. KRAUSS, G. "Influential Factors in the Fatigue Behavior of Case Hardened Steels Subjected to Rolling-Sliding Loads," Lecture to the Union of German Engineers, 3/22/72.
18. ALLSOPP, H.C., WEARE, A.T., and LOVE, R.J. "Resistance to Pitting of Gear Teeth, Re," Report 1959/2, Motor Industry Research Association of U.K., June, 1957.
19. LLEWELLYN, D.T. and COOK, W.T. "Metallurgy of Boron-Treated Low Alloy Steels," *Metals Technology*, December, 1974, p. 517.
20. GERMAN PATENT 1608732, "Procedure for Producing Tough Boron Treated Steels," September, 1969.

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