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Bending Fatigue Performance of Carburized Gear Steels

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ABSTRACT

This paper presents the results of a designed experiment and confirmational testing of the bending fatigue properties of boost-diffuse gas carburized gear steels for use in heavy duty truck gearing applications. Testing was conducted on simulated gear tooth samples in unidirectional four point bending under constant amplitude load control to crack initiation. The experiment was an L32 fractional factorial with eight levels of alloy grade, four levels of shot peening, and two levels each of grinding, case depth, surface carbon, ammonia additions, and tempering temperature. The SAE 4320 and boron containing alloy grades significantly outperformed the 8600 series steels. Surface conditions highly influenced the fatigue lives with low retained austenite, low surface carbon contents, and shallow intergranular oxidation depths resulting in superior performance. The beneficial effects of grinding and shot peening the tooth root following carburizing are discussed.

INTRODUCTION

The purpose of this report is to document the results of investigations into the effects of multiple material and process variables on the bending fatigue life of carburized gear steels. The gear steels evaluated in this study included many of the common SAE steels used in gearing: 8615; PS-15, an 8620 substitute; 8622; PS-16, an 8622 substitute; 8627; PS-18, an 8627 substitute; 4320; 94B17; and PS-19, a 94B17 substitute. Steel manufacturing variables included ingot cast steel, continuous cast steel, forging effects, and sample location. Heat treat variables included carburizing carbon potential, carburizing case depth, ammonia additions to the carburizing furnace, and tempering temperature. Post heat treat variables included shotpeening and grinding of the tooth root. All tests were run on a sample bar designed to simulate a single tooth from a forged, truck hypoid ring gear. The bar design will be discussed later. This study combined three project phases: preliminary testing to develop sample preparation expertise, a fractional factorial designed experiment, and confirmational testing to validate the designed experiment results.

EXPERIMENT DESIGN

The experiment was designed as a Taguchi L32 orthogonal array fractional factorial experiment. The experiment was saturated with main effect factors, providing negligible interaction observations. The purpose of the experiment was to provide a screening design of experiment, DOE, with less saturated DOE's to be persued in the near future to capture interactions. The main effect factors were 8 levels of alloy carburizing steel grades, 4 levels of shot peening, and 2 levels each of ammonia, grinding, total case depth, surface carbon content, and tempering temperature as shown in Table 1. One stress level, 1255 MPa, resulted in approximately equal numbers of low and high life results and was selected for all samples. The eight alloy steel grades covered a wide range of production available SAE gear steels including: 8615, PS-15, 8627, PS-18, 4320 at the low end of the hardenability band, 4320 at the high end of the hardenability band, 94B17, and PS-19.

The four levels of shot peening used, shown below, represent current gear peening practice. A compressed air type peening machine was used with the blast nozzle directed at the tooth root. One quarter of the samples, level 1, were tested in the unpeened condition while three quarters of the samples were peened under various conditions. Level 2 samples were peened with soft, 40 to 50 HRC, S-230 shot with a relatively high intensity of 0.47 mm Almen A arc height. Level 3 samples were peened with hard shot, 55 to 60 HRC, to a high intensity of 0.58 mm Almen A. Level 4 samples were peened to the same conditions as the Level 3 samples and followed up with a second pass of smaller diameter shot hard shot at a lower intensity to smooth the surface finish, as shown in Table A below.

Table A: Shot Peening Conditions

Shot Peen Level	Shot Size Diameter (mm)	Hardness (HRC)	Peen Intensity (mm Almen A)	Workpiece Peen Coverage (%)
1		No Peening		
2	S-230 (0.58)	40-50	0.47	300
3	S-230 (0.58)	55-60	0.58	300
4		Double Peen		
	S-230 (0.58)	55-60	0.58	300
	S-110 (0.28)	55-60	0.30	300

Metallurgical grade ammonia additions at 7% concentration were made to half of the samples during the 843 °C diffusion portion of the heat treat furnace carburizing cycle. This is a practice occasionally used to boost the surface hardness by forming iron nitrides through carbonitriding and by reducing the formation of bainite or pearlite at the surface grain boundaries caused by alloy loss as a result of intergranular oxidation, IGO, during carburizing.

Half of the samples were finish ground after carburizing in the tooth root prior to shot peening using a contoured wheel and flood cooling to simulate CBN, cubic boron nitride, grinding practice. Stock removal was 0.15 mm.

Two levels of total case depth, 1.8 mm and 2.5 mm were used to be typical of gearing case depths. Two levels of case carbon content, 0.70% and 1.00% carbon measured by spark emission spectrometer, were targeted during carburizing. A finished case carbon content of 0.70% is slightly below the eutectoid and primarily results in a Lathe type martensitic microstructure whereas 1.00% carbon is well above the eutectoid and results in primarily 100% Plate type martensite.

The last main effect factor was tempering temperature. Two levels of temperature, 93 °C and 204 °C, were used to temper the samples following carburizing. At 93 °C, the temperature is marginal for the precipitation of transition carbides whereas 204 °C results in significant carbide precipitation (1). All were tempered for 2 hours.

SAMPLE DESIGN AND PREPARATION

The bending fatigue properties were evaluated using unidirectional four point bending on simulated hypoid ring gear samples. The sample bars, shown in Figure 1, represent the idealized equivalent of a single tooth fatigue test sample cut from a finished ring gear, as shown in Figure 2. The notch dimensions and location are intended to approximate the stress concentration in the radius where the tooth dedendum blends into the tooth root. A typical stress concentration was calculated from a hypoid ring gear with a fillet radius at the base of the gear tooth of 2.8 mm, a pressure angle of 20°, and a working height of 11.2 mm which generates a stress concentration factor, Kt, of 1.632. The sample bars have been designed to provide a Kt = 1.628. The thickness and mass of the sample were chosen to approximate a truck hypoid ring gear, generating core hardnesses and case-to-core ratios indicative of heavy duty gears.

The samples were rough cut and Blanchard ground on all four surfaces with the sides finish ground to a 19.10 mm width. The samples were then 100% copper plated to 0.03 to 0.05 mm thickness. The final machining operation was to grind the top and bottom surfaces to remove the copper and finish to a height of 27.9 mm and contour grind the 7.6 mm radius notch to a finished depth of 6.4 mm. The surface finish in the direction of the notch was less than 0.9 micrometers on all as-ground samples measured. The copper plating on the side surfaces provides an effective stop-off to carburizing in order to: simulate a tooth cut from a finished ring gear, prevent through-carburizing of the sample corners to prevent alloy carbide formation and retained austenite in the corners, minimize premature crack initiation on the corners of the samples, and get uniform carbon penetration along the full length of the notch.

HEAT TREATMENT

All samples were boost-diffuse carburized. As the DOE required multiple combinations of carburized case depth, carbon potential, and ammonia level, the samples were split into 8 separate furnace loads in an in-out batch gas carburizing furnace. The samples were racked vertically in the baskets with a separate wire mesh screen near the top of the bars for additional support to minimize distortion. At least a 1 inch bar-to-bar spacing was maintained for uniform gas circulation and quenching. The samples were carburized at 927 °C at a carbon potential of 1.2% carbon, controlled with a calibrated oxygen sensor. Where ammonia additions were required, they were made during the 1550 °F diffusion cycle at a 7% concentration. None of the samples were grit blasted or shot cleaned following heat treatment to prevent additional compressive residual stresses from being induced into the samples. The DOE samples received two additional post-carburizing treatments.

The preliminary and confirmation test samples were carburized to two target case depth ranges, 1.5 mm and 2.3 mm total case depth. The 1.5 mm depth samples were all carburized in a pusher furnace on a 30 minute push cycle. The bars were racked and spaced identical to the previous samples. This boost-diffuse furnace cycle provided a 2 hour 871 °C preheat, 6 hours carburizing at 927 °C and 1.2% carbon potential, and 2 hours diffusion at 843 °C and 0.86% carbon potential. The oil quench temperature was 52 to 66 °C with both agitators at high speed. The samples were tempered for 2 hours at 177 °C. The 2.3 mm depth samples were carburized in the same furnace and at the same carbon potentials but on a longer, 60 minute, push cycle.

FATIGUE TECHNIQUE

All samples were tested to crack initiation and full fracture on closed loop electrohydraulic fatigue testers under load controlled, constant amplitude, unidirectional four point bending with the load application points P1 and P2 and reaction points R1 and R2, as shown in Figure 1. All samples were tested at 10 Hertz with a sinusoidal load wave. This frequency held the maximum temperature rise in the samples due to straining to less than or equal to 4 °C. Crack initiation was detected by monitoring deflection changes beyond the initial sample deflection. Visual checks and magnetic particle inspection were used to confirm the presence of a crack whenever the deflection limit was tripped. The maximum compressive load was calculated based on the stress required, individual sample dimensions, and the stress concentration factor. In all cases the minimum compressive load was 45 kilograms or approximately 10 MPa to insure retention of the sample in the four point bending test fixture during minimum loading. The stress ratio, R, was 0.01 for all samples. The preliminary and confirmational samples were tested at two target stress levels, 951 MPa and 1255 MPa maximum bending stress in the root of the notch with the stress concentration factor included. The designed experiment samples were tested only at 1255 MPa in order to minimize the number of runout samples as well as minimize the number of samples with low lives to failure. All samples were tested to failure or 1 to 2 million cycles. Each runout sample from the designed experiment was repeated to verify the runout data point. Following the detection of crack initiation, the number of cycles was recorded and the sample was restarted under load control at the same load level used to initiate the crack. The number of cycles to full fracture was recorded. The load cell shunt calibration was checked and recorded before each sample to verify cell calibration. Sample condition and test observations were recorded for each bar.

PRELIMINARY TESTING RESULTS

The purpose of these preliminary test sequences was to develop the expertise necessary to properly select the location and position of the sample bars. This testing investigated four aspects of sample preparation: differences between bar stock and forged stock generated samples, the effects of sampling from various locations within the bar stock, the effects of sampling from various bar stock sizes, and the difference between ingot cast and continuous cast steels. The sampling locations and preliminary, confirmation, and designed experiment samples are shown in Table 2.

Generating samples from bar stock is more economical and expedient than using gear forgings but the possibility that samples cut from bar stock could have different properties than those cut from forged gear blanks had to be investigated. In all cases the samples from bar stock were cut longitudinal to the rolling direction of the bar. The samples were cut with the notch toward the outside of the bar stock and transverse to the rolling direction. Samples cut from unmachined ring gear forgings were positioned with the root of the notch at approximately the same depth as a production cut gear tooth root. As shown in Table 3, the fatigue results of bar stock generated samples were compared to samples cut from forged ring gear blanks using the same heat of steel for four steel grades; SAE 8622, PS-16, 94B17, and PS-19. The sample populations were compared to determine what statistical differences were present using a Nonparametric statistical approach. This statistical program was developed by Rockwell Automotive Operations staff statisticians in recognition of the non-normality of fatigue test data. This program does not assign a distribution, unlike the Weibull technique, and is ideal for small sample sizes, on the order of one to six samples of each material and process variation. Only data that demonstrated a significant difference at 5 or 10% risk, equivalent to 90 and 95% confidence, was considered for detailed analysis. The results, shown in Table 3, show no major trend for forged samples to have significantly different crack initiation bending fatigue lives than bar stock samples at a 5 or 10% risk level based on 32 groups of samples or a total of 145 samples. This greatly simplified all future sample preparation provided the gear design being modeled has cut teeth. Gears with forged-in teeth, in which the forging flow lines are not disrupted by machining, could demonstrate better lives than cut gears.

Next, samples were prepared from one ingot cast heat of SAE 8622 and one continuous or strand cast heat of SAE 8622 to evaluate the influence of steel manufacturing technique on bending fatigue life, especially considering the trend for ingot cast steel mills to be replaced by continuous casting mills in increasing numbers in world steel production. The continuous cast steel billet used in this study received a 4.9:1 rolling reduction. As shown in Table 4, there was no consistent significant difference with respect to crack initiation bending fatigue life between ingot and continuous cast steel at reductions greater than or equal to 4.9:1 based on 10 groups of samples or 45 total samples (2).

Next, samples cut from the half radius and from the O.D. of 140 mm and 152 mm continuous cast bar stock were compared to determine if depth is critical, particularly in continuous cast steel where centerline porosity from the casting operation may adversely affect the fatigue life on samples cut from deeper sections of the bar stock. Samples cut from the one-half radius were positioned with the root of the notch at the half radius. As shown in Table 5 for SAE 8622 and PS-16, the results were mixed but in general there was no discernible trend based on 12 groups of samples or 58 total samples.

Lastly, samples from the O.D. of a 59 mm diameter round and the O.D. of a 140 mm diameter round of SAE PS-19 from the same heat were compared. In general there was no significant statistical difference based on 8 groups of samples or 36 total samples, as shown in Table 6.

In general, the preliminary testing showed that future sample preparation could be greatly simplified permitting samples to be taken from any location in ingot cast or strand cast bar, except the very center where centerline shrink and inclusion levels could be excessive. These conclusions were incorporated into the designed experiment sample preparation.

DESIGNED EXPERIMENT RESULTS AND DISCUSSION

The analysis of variance for the DOE found the main effect factors to influence the bending fatigue lives in the following order, from most important to least important: shot peening level, SAE alloy grade, grinding, carbon content in the carburized case, tempering temperature, ammonia additions, and finally, total case depth. The sample test results are compiled in Table 7.

The effect of shot peening on the overall B50 life of the samples is shown in Table B below. Level 4 increased the effective life to crack

Table B: Shot Peening LIfe Effects

		B50 Cycles to Failure	Factor
Level 1	No peening	12 303	1.0x
2	Soft shot	4 074	0.3x
3	Hard shot	16 982	1.4x
4	Double peen hard shot	107 152	8.7x

initiation by a factor of nearly 9 compared to samples without peening (3)(4). Shotpeen levels 2 and 3 were more representative of typical shot peening practice where only one size shot is used. As shown, level 3 provided only a 38% increase in life while level 2 resulted in a decrease in fatigue life. A decrease is surprising but may be partially explained later.

Peening was found to have three distinct effects on the samples; increased compressive stress, decreased retained austenite contents, and increased surface roughness. As expected, peening significantly increased the compressive residual stress, as measured on the surface by Fastress x-ray diffraction. Level 4 induced 3 times the residual stress compared to samples without peening (5). As shown in Table D below, the soft shot used in the level 2 samples provided no effective reduction in the retained austenite level, with the level 3 and level 4 hard shot providing 50% reductions. The effect of shot peening on the surface finish is shown in Table E below. The decrease in surface finish for level 4 is the result of the second pass using small diameter shot.

Table C: Residual Stress Levels

	Average Residual Stress (MPa)
Level 1	-249
2	-459
3	-555
4	-840

Table D: Effect of Peening on Retained Austenite Level

	Average Retained Austenite by X-Ray Surface, Post-Peen	% Decrease
Level 1	16.2%	--
2	17.0%	0%
3	8.4%	48%
4	6.7%	59%

Table E: Effect of Peening on Surface Finish

	Average Surface Finish (micrometers)
Level 1	0.56
2	0.76
3	1.12
4	0.97

The decrease in fatigue life observed for level 2 is believed to be the result of the shot increasing the surface finish but failing to reduce the retained austenite levels while only providing a moderate increase in residual stress and a much shallower stress pattern. Level 4, demonstrating the highest fatigue life, provided maximum residual stress, a maximum reduction in retained austenite, and improved surface finish. It is interesting to note in Table 7 that long bending fatigue lives can be achieved without the expense of shot peening, provided the right combination of conditions are maintained. Of the ten samples generating fatigue lives greater than 580 kilocycles at 1255 MPa, six samples were not peened.

The second most important factor was the SAE alloy grade. As shown in Table F below, the two SAE 4320 alloys were the top performers of the eight steel grades tested. The second top performer was the SAE 8627 substitute, PS-18. This is a surprising result considering the poor performance of 8627, the high nickel version of PS-18. The next group of performers were the boron steels, 94B17 and PS-19. As shown, 94B17 and PS-19, the low nickel substitute for 94B17, were equivalent performers. The worst performers were the 8600 series steels which included PS-15, 8615, and 8627.

Table F: Alloy Grade Ranking

	Alloy	B50 Cycles to Failure	Factor
Level 5	4320 low	102 329	15.1x
1	4320 high	36 308	5.4x
2	PS-18	25 119	3.7x
4	PS-19	13 804	2.0x
8	94B17	13 490	2.0x
6	PS-15	9 550	1.4x
7	8615	7 586	1.1x
3	8627	6 761	1.0x

The third most important factor was grinding, with 0.15 mm removal out of the notch root resulting in a 6 fold increase in the average bending fatigue life compared to no grinding, as shown in Table G below. Though the ground surface finishes were the same as the carburized finishes, the influence of grinding appears to be the removal of the oxide films and the intergranular oxidation associated with the carburizing furnace atmospheres, factors that will be further discussed later (6).

Table G: Effect of Grinding on Life

		B50 Cycles to Failure	Factor
Level 1	Grind 0.15 mm	44 157	6.4x
2	No grind	6 918	1.0x

The fourth most important factor was the surface carbon level generated by the carbon potential of the carburizing furnace. As shown in Table H below the bending fatigue life increased five fold when the case carbon

Table H: Effect of Surface Carbon on Life

		B50 Cycles to Failure	Factor
Level 1	Low carbon 0.70%	38 905	5.0x
2	high carbon 1.00%	7 762	1.0x

level was held to the low range, which varied from 0.60 to 0.92% carbon. This is believed to be due to the formation of larger amounts of plate martensite at the higher carbon contents. Considerable work has shown that microcracks are associated with plate martensite formations and contribute to premature crack initiation (7). All of the runout samples had surface carbon contents between 0.60 and 0.85% carbon.

The last three factors, ammonia additions, tempering temperature, and case depth provided weaker conclusions of a statistically lower significance, however the results were still interesting and somewhat logical. In general, no ammonia was preferable to 7% ammonia with the life of the no ammonia samples approximately 3 times that of the ammonia treated samples. This is a logical conclusion considering that carbonitrided parts are typically more brittle and lower toughness than carburized parts. The effect of nitrogen on lowering the martensite start, Ms, is shown by the 50% increase in retained austenite in the ammonia treated samples. The 204 °C temper for two hours was preferable to 93 °C, as shown in Table I below. This is logical in that the

Table I: Effect of Variables on Life

		B50 Cycles to Failure	Factor	% Retained Austenite
Level 1	7% ammonia	10 902	1.0x	21.9
2	0% ammonia	27 742	2.5x	14.4
1	93 °C temper	10 871	1.0x	
2	204 °C temper	27 821	2.6x	
1	Low case depth 1.8 mm	23 988	1.9x	
2	High case depth 2.3 mm	12 589	1.0x	

transition epsilon carbides precipitated at 204 °C increase the toughness and reduce the adversely high residual stresses from quenching. In the weakest conclusion, total case depths in the range of 1.8 mm were found to be preferable to 2.5 mm. Though deeper case depths result in lower compressive stresses, confirmation testing over a larger group of samples is required to verify such a weak conclusion.

CONFIRMATION TESTING

Confirmation testing with larger sample sizes is necessary to validate the results of any designed experiment. Of the seven main effect factors, three factors, alloy steel grade, case carbon content, and total case depth were selected in six different alloy grades for confirmation. The alloy grades, shown in Table 2, were SAE 8622, PS-16, 8627, 94B17, PS-19, and a midrange hardenability SAE 4320. SAE 4320 was not retested in detail because of it's overwhelming influence in the designed experiment and the selection of alloys was narrowed in hardenability range to further challenge the designed experiment results. Multiple heats of SAE 8622, 8627, and 94B17 were included. The heats were the same as those tested in the preliminary tests. PS-16 and SAE 8622 were substituted for the PS-15 used in the designed experiment because samples were available from the preliminary tests. Shot peening and grinding effects have not been confirmed at this time.

Total case depth, the weakest main effect factor in the designed experiment, was further tested because of it's expensive influence on carburizing furnace time. Two total case depths, 1.5 mm and 2.3 mm, were examined in detail. Controlling the total or effective case depths when multiple steel grades are run in the same furnace batch can be accomplished only by aiming for the median case depth and accepting the variations that occur. Among the eight heats of steel carburized, the resulting case depths were extremely close to the target values. The median case depth for the 1.5 mm case was equal to 1.60 mm with one standard deviation equal to 0.11 mm. The median case depth for the 2.3 mm case was equal to 2.29 mm with one standard deviation equal to 0.22 mm. The sample data provided extensive comparisons between the low and high case depths and, as shown in Table 8, total case depth variations from 1.4 mm to 2.6 inch depth had no effect on the bending fatigue lives, based on 48 groups of samples, 216 total samples. This confirms the designed experiment results.

The next main effect factor to be confirmed was the effect of alloy grade, found to be a strong factor in the designed experiment. The first interesting feature to be noted in Table 9 is that the low nickel substitutes performed comparable to the equivalent nickel grades. There was no significant difference at 10% risk between SAE 8622 and PS-16, the low nickel substitute for 8622, based on 16 groups of samples, 72 total samples. The current cost savings in substituting PS-16 for 8622 is approximately 8%. Similarly there was no significant difference between 94B17 and PS-19, a low nickel substitute for 94B17, based on 14 groups of samples, 65 total samples. This confirms the results of the designed experiment on the two boron steel grades. The current cost savings in substituting PS-19 for 94B17 is approximately 6%.

As in the designed experiment, the boron containing steels, SAE 94B17 and PS-19 were the top performing steels of the five steel grades tested, as shown in Table J below. PS-19 and SAE 94B17 were 4.1 and 3.3 times better than SAE 8622 at the designed experiment stress level of 1255 MPa and 5.8 and 4.6 times better at the lower stress level of 951 MPa (8).

Table J: Alloy Grade Ranking

Alloy	B50 Cycles to Failure 1255 MPa	Factor	951 MPa	Number of Samples
PS-19	34 300	4.1x	331 600	50
94B17	27 400	3.3x	250 200	50
8627	14 300	1.7x	50 500	22
PS-16	10 100	1.2x	76 500	36
8622	8 400	1.0x	57 400	66

Currently PS-19 is 5% lower in cost than 8627 and 3% higher than PS-16. SAE 8627 continued to perform erratically and may be due to bainitic transformations (9). Though the performance is improved over that in the designed experiment, the performance at 951 MPa is quite poor. The improved performance of all five steels compared to the designed experiment is due to the lack of ammonia additions and more moderate tempering temperatures and case carbon levels. The tempering temperature was 177 ° C and the average carbon content was 0.86%.

The last result of the designed experiment to be confirmed was case carbon content, a moderately strong effect. SAE 4320 samples carburized to a target carbon level of 0.7%, compared to samples carburized to a target level of 1.0% carbon were found to have an average life 5.3, 8.7, and 18.5 times greater than the life of the high carbon samples at 1034 MPa, 1200 MPa, and 1324 MPa stress levels, respectively. Compared to the 5 fold increase found for all steel grades in the designed experiment, this indicates that SAE 4320 may be more sensitive to carbon content, particularly at higher stress levels.

Table K: Effect of Surface Carbon on Life

	B50 Cycles to Failure 1034 MPa	1200 MPa	1324 MPa	Number of Samples
Low Carbon 0.70%	330 000	63 700	62 400	21
High Carbon 1.00%	62 900	7 300	3 400	12

MULTIPLE REGRESSION ANALYSIS

In addition to the main effect factors controlled in the designed experiment, many other uncontrolled factors were collected on the DOE samples. They included; surface hardness, core hardness, core and case hardenability per SAE J406, core chemical analysis, effective case depth, depth of intergranular oxides at the surface grain boundaries, prior austenitic grain size, visual retained austenite levels below the surface of the notch, and the internal sulfide and oxide contents measured per SAE J422. Several other factors including notch surface finish, notch transverse residual stress measured by the Fastress technique, and notch surface retained austenite measured by X-ray diffraction have been previously discussed. These factors were correlated to the log 10 bending fatigue cycle lives to determine if and to what extent they may have influenced the lives observed. Because of their high degree of interaction, these variables were entered into a stepwise multiple regression statistical routine that regresses the data against the logarithmic cycle lives and builds a model from the statistically significant data while rejecting the insignificant data. The alloy grades were entered as numbers one through eight corresponding to the levels shown in Table 1.

Five of the variables were found to have F ratios that were significant and together they explained 70% of the variations in cycle lives observed. They included IGO depth, grain size, SAE alloy grade, retained austenite level, and case hardenability. All of the other variables were found to be less significant in comparison. Bending fatigue life was found to increase as retained austenite decreased, IGO depth decreased, the grain size increased, and case hardenability decreased.

Table L: Regression Analysis

Model Variables	Coefficient	Sum of Squares	DF	F Ratio	p value
Constant	14.925				
Retained Austenite	-0.0459	3.802	1	7.23*	0.0116
Surface Finish	0.0197	.487	1	.93	0.3537
SAE Alloy Grade	-0.2037	5.357	1	10.18*	0.0033
IGO Depth	-0.0013	8.371	1	15.91*	0.0004
Grain Size Number	-0.7688	13.136	1	24.97*	0.0000
Case Hardenability	-0.3989	5.003	1	9.51*	0.0044

$F_{0.95}(1,6) = 5.99$; *Any F ratio >5.99 is significant at the 95% confidence level

With the results of this regression it is possible to better understand the designed experiment results. Despite shot peening's beneficial effects on residual stress, the main factor appears to be shot peening's ability to reduce the retained austenite levels. In mechanically subjecting the gear surface to plastic, permanent deformation in order to generate residual stresses, peening converts the thermally and mechanically unstable microstructural constituent, austenite, retained from the high temperature portions of the carburizing operation, into stable martensite. Generating approximately a 4% volume expansion during the conversion, the martensite further strains the atomic lattice, generating additional compressive stresses. Retained austenite, on the other hand, is very weak and rapidly initiates microcracks during bending fatigue operation. Though, over time in field service, thermal excursions and mechanical working will convert some retained austenite to martensite, this requires considerable time and puts the gear at risk for premature failure. During all of these tests the levels of retained austenite did not change as a result of the testing, demonstrating the relative mechanical stability of austenite. The proper level of retained austenite in carburized microstructures continues to be controversial with zero retained austenite required in aircraft gearing where dimensional stability and long life are paramount, to 50% in roller bearings where retained austenite improves the contact fatigue life (10)(11)(12). This study suggests that retained austenite should be maintained as low as possible and should not exceed 22% for improved bending fatigue for gearing applications where considerable sliding occurs, as in hypoid gearing. The lower B50 cycle lives to crack initiation in the DOE compared to the preliminary and confirmation tests appear to be the result of the use of higher levels of retained austenite, as well as higher case carbon contents, the use of ammonia, and lower tempering temperatures discussed earlier.

The regression suggests that gear steels should be selected for optimum bending fatigue performance with lower core hardenabilities and carburized at lower carbon potentials, i.e. less than 0.85% carbon to reduce the case hardenabilities. The three fold improvement in low hardenability SAE 4320 compared to high hardenability 4320 shows the potential effect of hardenability. Lower carbon potentials also result in lower retained austenite levels.

The regression also indicates that intergranular oxidation should be minimized. This can be accomplished by grinding the tooth root following carburizing. Grinding also removes retained austenite and exposes lower carbon content or lower case hardenability subsurface layers, explaining why grinding was within the top three important factors in the DOE. Grinding, except to correct rough surface finishes, may not be neccesary to improve gear life provided reasonable retained austenite and carbon potentials are maintained. IGO depth can also be minimized by the use of vacuum carburizing instead of gas carburizing and selecting alloys with elemental analyses that minimize the use of high oxidizing potential elements. Alloying elements such as nickel and molybdenum should be used preferential to manganese and chromium. Aluminum, silicon, and phosphorus contents should be minimized. Boron has a high oxidation potential but is used in extremely small quantities, on the order of 0.0005%.

Conventional wisdom indicates that bending fatigue life should increase as the grain size number increases or grain diameter decreases within any one steel alloy because finer grain results in more notch toughness, higher tensile and yield strength, lower internal stress after hardening, less retained austenite, and lower hardenability (13). This study indicates that coarser grain size is better in carburizing steels in the range of 9 to 11 grain size, but further work is required over a broader range of grain sizes to confirm the effect of grain size.

As expected, the factors that related only to subsurface conditions, such as core hardness, subsurface retained austenite content, inclusion counts, etc., had no influence on the bending fatigue lives. Subsurface retained austenite measured visually on a metallograph was found to be an extremely poor indicator of the amount of surface retained austenite measured by X-ray diffraction and thus correlated poorly to the bending fatigue lives. Though sulfide inclusion levels varied from 1 to 5 per SAE J422, the decision to orientate the samples longitudinal to the rolling direction does minimize the effect of sulfide and oxide inclusions. Surface inclusions parallel to the tooth root can be expected to adversely affect bending fatigue life and should be considered whenever sample preparation is planned. Inclusions in ring or pinion gear forgings can be expected to lie in many orientations. Similarly, no one chemical element including nickel correlated to the bending fatigue lives, though it is possible that there is a synergistic effect of several elements.

CONCLUSIONS

1. The longest bending fatigue life to crack initiation was observed in SAE 4320 steel, followed by PS-18 and the boron steels, 94B17 and PS-19. The worst performers were the 8600 series and 8600 series substitute steels including 8615, PS-15, 8622, and PS-16, with all providing nearly equivalent performance. SAE 8627 provided erratic results, performing poorly in the designed experiment at 1255 MPa and poorly at 951 MPa in the confirmational testing. SAE 4320 demonstrated an average 5 fold improvement over the boron steels at 1255 MPa bending stress based on the DOE results, but the cost premium is 35%. SAE 94B17 and PS-19 demonstrated an average 3 fold improvement over the 8600 series at 1255 MPa and an average 5 fold improvement at 951 MPa based on the confirmation test results. PS-19 is 5% lower in cost than 8627 and 3% higher than the least expensive 8600 grade substitutes.

2. Surface conditions highly influence the bending fatigue life. Long bending fatigue lives can be achieved in all steels including the 8600 series, without shot peening, provided surface retained austenite was maintained to as low a level as possible, surface carbon content was held below 0.85% carbon, surface finish was held to below 0.8 micrometers, and intergranular oxidation was minimized. Grinding the tooth root had a significant influence on bending fatigue life, increasing the life by a factor of 6 compared to no grinding. It can improve the surface finish, remove intergranular oxidation, retained austenite, and high surface carbon layers, but equivalent results could also be achieved by following the recommendations above. Alloys that minimize the use of high oxidizing potential elements such as manganese, chromium, aluminum, silicon, and phosphorus, in favor of elements such as nickel and molybdenum may offer higher bending fatigue lives. Factors that relate only to subsurface conditions such as core hardness and subsurface retained austenite had no significant influence on bending fatigue life.

3. Total carburized case depths varying from 1.4 mm to 2.6 mm had no effect on the bending fatigue performance at 951 and 1255 MPa.

4. Typical gear shot peening practice using large diameter, hard shot provided a 38% increase in fatigue life at 1255 MPa. Double shot peening with hard shot resulted in major improvements in bending fatigue life, on the order of 9 fold increases, but the cost for double peening is

quite high compared to typical practice. Double peening requires total coverage with large diameter, hard shot followed by small diameter shot to improve the surface finish. Two set-ups are required. Shot peening's primary influence in improving the bending fatigue life was in reducing the retained austenite levels, with compressive residual stress improvements as a secondary effect. Shot peening was not mandatory for long bending fatigue life but improved the probability of long life. Typical gear shot peening practice using soft shot provided no overall improvement except when the retained austenite content was low.

5. The low nickel substitutes for SAE 94B17 and SAE 8622 steel were found to be equivalent in performance to their high nickel versions.

6. Ammonia additions during the diffuse portion of the carburizing cycle and low tempering temperatures, i.e. 93 °C, adversely affected the bending fatigue life.

7. Ingot cast steel and continuous cast steel within the same alloy grade were found to have equivalent crack initiation bending fatigue lives at 951 MPa and 1255 MPa stress using strand rolling reduction ratios greater than or equal to 4.9:1.

8. This testing reports on only bending fatigue, one of several types of gear distress that can be encountered in service. Overload that results in case crushing and impact loading were not simulated by this testing. At lower loads, contact or pitting fatigue becomes a significant factor, though the life to failure is extended by the lower loads. Abrasive wear, due to contamination of the oil by dirt or wear particles, and adhesive wear or scoring due to the use of incorrect hypoid lubes, water contaminated lubes, or low lube levels are encountered less often but are still significant factors. The incidence of adhesive wear has been greatly reduced by the widespread adoption of GL-5 rated gear lubes. Recommendations as a result of this testing focus only on improving the bending fatigue performance and may have beneficial or adverse effects under contact, impact, or case crushing service conditions.

ACKNOWLEDGEMENT

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TABLE 1
DESIGNED EXPERIMENT
FACTORS, LEVEL DEFINITION, LAYOUT

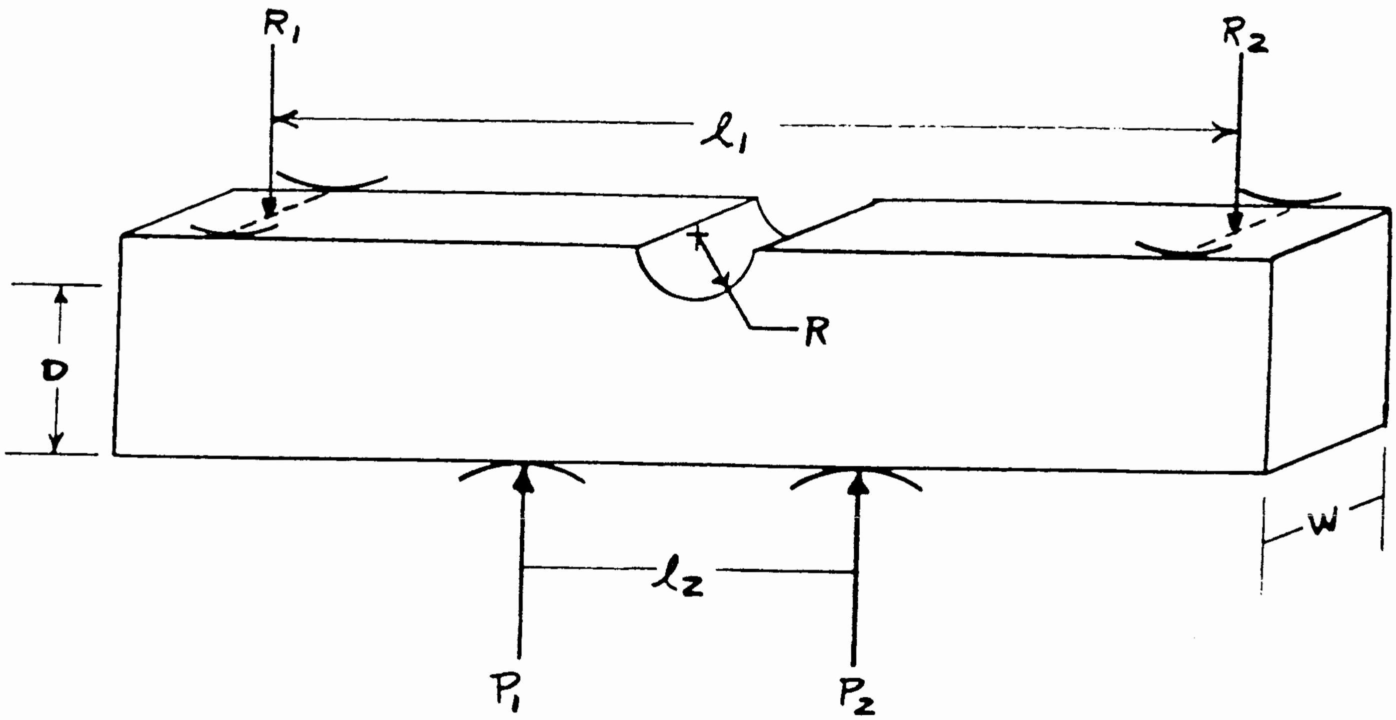
FACTORS	LEVEL 1	LEVEL 2	LEVEL 3	LEVEL 4	LEVEL 5
SAE ALLOY GRADE	4320 HIGH	PS-18	8627	PS-19	4320 LOW
ALLOY CODE	B	G	I	L	C
			LEVEL 6	LEVEL 7	LEVEL 8
SAE ALLOY GRADE			PS-15	8615	94B17
ALLOY CODE			K	J	F
AMMONIA	7%	0%			
FINAL GRINDING	0.15 mm	NO GRIND			
SHOT PEENING	NO PEEN	ONCE SOFT	ONCE HARD	TWICE HARD	
TOTAL CASE DEPTH	1.8 mm	2.5 mm			
SURFACE CARBON	0.70%	1.00%			
TEMPERING TEMPERATURE	93°C	204°C			

TRIAL NUMBER	ALLOY GRADE LEVEL	AMMONIA LEVEL	GRIND LEVEL	PEEN LEVEL	CASE DEPTH LEVEL	SURFACE CARBON LEVEL	TEMPER TEMP LEVEL
1	1	1	1	1	1	1	1
2	1	1	2	2	2	2	1
3	1	2	1	3	1	1	2
4	1	2	2	4	2	2	2
5	2	2	1	1	2	1	1
6	2	2	2	2	1	2	1
7	2	1	1	3	2	1	2
8	2	1	2	4	1	2	2
9	3	2	2	1	2	2	2
10	3	2	1	2	1	1	2
11	3	1	2	3	2	2	1
12	3	1	1	4	1	1	1
13	4	1	2	1	1	2	2
14	4	1	1	2	2	1	2
15	4	2	2	3	1	2	1
16	4	2	1	4	2	1	1
17	5	2	1	1	1	2	1
18	5	2	2	2	2	1	1
19	5	1	1	3	1	2	2
20	5	1	2	4	2	1	2
21	6	1	1	1	2	2	1
22	6	1	2	2	1	1	1
23	6	2	1	3	2	2	2
24	6	2	2	4	1	1	2
25	7	1	2	1	2	1	2
26	7	1	1	2	1	2	2
27	7	2	2	3	2	1	1
28	7	2	1	4	1	2	1
29	8	2	2	1	1	1	2
30	8	2	1	2	2	2	2
31	8	1	2	3	1	1	1
32	8	1	1	4	2	2	1

FIGURE 1

FOUR POINT BENDING FATIGUE

TEST BAR



L = 6 ± .030 INCH
H = 1.100 ± .001 INCH
D = 0.850 ± .001 INCH
W = 0.750 ± .001 INCH
R = 0.300 ± .001 INCH
ALL SIDES SQUARE WITHIN
0.0005 INCH

FIGURE 2
IDEALIZED LOCATION
OF SINGLE TOOTH BENDING
FATIGUE SAMPLE

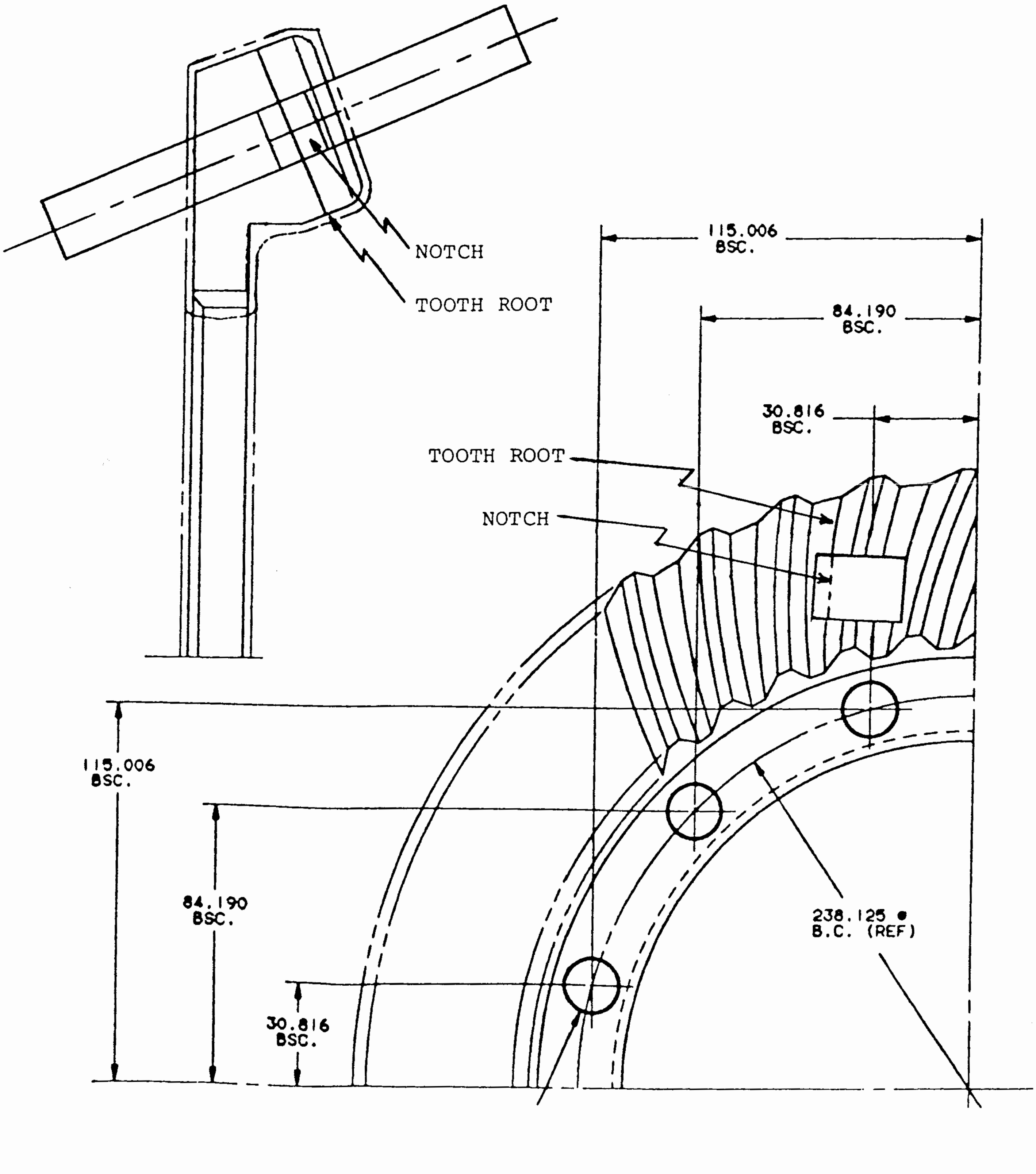


TABLE 2

STEEL DATA AND SAMPLE LOCATIONS

SAE GRADE	ALLOY CODES	STEEL SOURCE	HEAT NUMBER	STEEL PRACTICE INGOT CAST	STEEL PRACTICE CONTINUOUS CAST	BAR STOCK SIZE	SAMPLE LOCATION ONE HALF RADIUS	SAMPLE LOCATION O.D.	SAMPLE LOCATION RING GEAR FORGING	SAMPLE LOCATION THROUGH-OUT SECTION	SAMPLES PER CROSS SECTION
					PRELIMINARY & CONFIRMATION SAMPLES						
8622	L,LX	SOURCE 1	A	X		140 mm RCS		X			10
8622	LF,LFX	SOURCE 1	A	X		FORGING			X		5/FORGING
8622	P	SOURCE 2	B		X	140 mm RCS		X			10
8622	P1	SOURCE 2	B		X	140 mm RCS	X				8
8622	PF,PFX	SOURCE 2	B		X	FORGING			X		5/FORGING
PS-16	U,UX	SOURCE 3	C		X	152 mm RCS		X			16
PS-16	U1,U1X	SOURCE 3	C		X	152 mm RCS	X				10
PS-16	UF,UFX	SOURCE 3	C		X	FORGING			X		5/FORGING
8627	T,TX	SOURCE 4	D	X		59 mm RND		X			2
8627	B2,B2X	SOURCE 5	E	X		102 mm RCS		X			8
94B17	N,NX	SOURCE 6	F	X		140 mm RCS		X			10
94B17	NF,NFX	SOURCE 6	F	X		FORGING			X		5/FORGING
94B17	S,SX	SOURCE 6	G	X		59 mm RND		X			2
PS-19	R,RX	SOURCE 4	H	X		59 mm RND		X			2
PS-19	M,MX	SOURCE 4	H	X		140 mm RCS		X			12
PS-19	MF,MFX	SOURCE 4	H	X		FORGING			X		5/FORGING
4320 MID HARD	F2	SOURCE 12	Q	X		140 mm RCS				X	12
					DESIGNED EXPERIMENT SAMPLES						
8615	J	SOURCE 7	I	X		165 mm RCS				X	24
PS-15	K	SOURCE 7	J	X		140 mm RCS				X	15
8627	I	SOURCE 7	K	X		89 mm RND				X	4
PS-18	G	SOURCE 8	L	X		89 mm RND				X	4
4320 LOW HARD	C	SOURCE 9	M	X		92 mm RND				X	4
4320 HIGH HARD	B	SOURCE 10	N	X		152 mm RCS				X	24
94B17	F	SOURCE 11	O	X		67 mm RND				X	2
PS-19	L	SOURCE 4	P	X		73 mm RND				X	2

1. ALL SAMPLES CUT LONGITUDINAL TO ROLLING DIRECTION
2. O.D. SAMPLES CUT WITH NOTCH TO THE O.D.
3. ONE HALF RADIUS SAMPLES CUT WITH THE NOTCH ROOT AT ONE HALF RADIUS DEPTH
4. RING GEAR FORGING SAMPLES CUT WITH NOTCH AT APPROX THE SAME DEPTH AND ORIENTATION AS GEAR TOOTH ROOT
5. NONE OF THE SAMPLES WERE GRIT BLASTED FOLLOWING CARBURIZING
6. ALL SAMPLES CODED WITH X SUFFIX WERE CARBURIZED TO 1.5 mm TOTAL CASE DEPTH; ALL OTHERS 2.3 mm ON PRELIMINARY AND CONFIRMATION SAMPLES

CHEMICAL ANALYSES (PERCENT)	C	P	Si	S	Mn	Cr	Mo	Ni	B	Al	Cu
L,LX,LF,LFX	0.21	0.011	0.28	0.035	0.80	0.60	0.17	0.45	0.0000		
P,P1,PF,PFX	0.23	0.017	0.26	0.023	0.79	0.50	0.18	0.59	0.0000	0.03	0.09
U,UX,U1,U1X,UF,UFX	0.23	0.010	0.25	0.032	1.01	0.55	0.19	0.12	0.0000		0.22
T,TX	0.28	0.014	0.25	0.020	0.85	0.59	0.16	0.42	0.0000		0.15
B2,B2X	0.27	0.010	0.22	0.020	0.87	0.50	0.17	0.48	0.0000		
N,NX,NF,NFX	0.17	0.007	0.25	0.025	0.84	0.37	0.09	0.35	0.0000		
S,SX	0.18	0.006	0.24	0.019	0.83	0.37	0.14	0.41	0.0020		
R,RX,M,MX,MF,MFX	0.20	0.013	0.28	0.022	1.10	0.53	0.10		0.0012		
F2	0.22	0.025	0.19	0.027	0.61	0.57	0.23	1.79	0.0000	0.04	0.05
J	0.18	0.014	0.24	0.016	0.74	0.55	0.21	0.44	0.0000	0.02	0.15
K	0.22	0.015	0.30	0.011	0.92	0.50	0.19	0.19	0.0000	0.01	0.12
I	0.28	0.016	0.32	0.023	0.87	0.51	0.19	0.54	0.0000	0.04	0.21
G	0.30	0.019	0.28	0.029	1.06	0.54	0.17	0.11	0.0000	0.02	0.15
C	0.16	0.010	0.19	0.025	0.42	0.43	0.21	1.60	0.0000	0.01	0.02
B	0.22	0.023	0.24	0.026	0.61	0.52	0.29	1.56	0.0000	0.02	0.01
F	0.19	0.018	0.25	0.014	0.77	0.31	0.11	0.23	0.0005	0.01	0.10
L	0.20	0.021	0.25	0.033	1.06	0.49	0.15	0.17	0.0005	0.03	0.19

920533

TABLE 3
FORGED & ROLLED SAMPLE COMPARISON
CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE FORGED	SAMPLE CODE ROLLED	STRESS LEVEL (MPa)	CONSTANTS			SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) FORGED	B50 LIFE (CYCLES) ROLLED	NUMBER OF SAMPLES FORGED	NUMBER OF SAMPLES ROLLED
LFX	LX	951	1.5 mm	INGOT	8622	SIGNIFICANT DIFFERENCE AT 5% RISK	94 400	54 000	5	5
LF	L	951	2.3 mm	INGOT	8622	SIGNIFICANT DIFFERENCE AT 25% RISK	44 400	66 200	5	4
LF	L	1255	2.3 mm	INGOT	8622	NO SIGNIFICANT DIFFERENCE AT 5% RISK	6 610	7 540	4	4
PF	P	951	2.3 mm	STRAND	8622	SIGNIFICANT DIFFERENCE AT 5% RISK	76 400	28 800	3	5
PF	P	1255	2.3 mm	STRAND	8622	SIGNIFICANT DIFFERENCE AT 25% RISK, NONE AT 5%	8 590	7 120	4	6
UFX	UX	951	1.5 mm	STRAND	PS16	SIGNIFICANT DIFFERENCE AT 10% RISK	90 200	54 700	5	6
UFX	UX	1255	1.5 mm	STRAND	PS16	NO SIGNIFICANT DIFFERENCE AT 5% RISK	12 700	12 700	4	4
UF	U	951	2.3 mm	STRAND	PS16	NO SIGNIFICANT DIFFERENCE AT 5% RISK	86 200	80 300	4	4
UF	U	1255	2.3 mm	STRAND	PS16	SIGNIFICANT DIFFERENCE AT 25% RISK	12 000	8 540	4	5
NFX	NX	951	1.5 mm	INGOT	94B17	SIGNIFICANT DIFFERENCE AT 10% RISK	>2 000 000	110 000	5	4
NFX	NX	1255	1.5 mm	INGOT	94B17	SIGNIFICANT DIFFERENCE AT 25% RISK	31 800	24 000	5	3
NF	N	951	2.3 mm	INGOT	94B17	SIGNIFICANT DIFFERENCE AT 25% RISK	355 000	107 000	5	4
NF	N	1255	2.3 mm	INGOT	94B17	SIGNIFICANT DIFFERENCE AT 25% RISK, NONE AT 5%	23 800	19 200	5	5
MFX	MX	951	1.5 mm	INGOT	PS19	NO SIGNIFICANT DIFFERENCE AT 25% RISK	266 000	339 000	5	6
MF	M	951	2.3 mm	INGOT	PS19	SIGNIFICANT DIFFERENCE AT 10% RISK	314 000	106 000	5	4
MF	M	1255	2.3 mm	INGOT	PS19	NO SIGNIFICANT DIFFERENCE AT 25% RISK	25 500	21 700	4	4

TABLE 4
INGOT CAST STEEL & STRAND CAST STEEL COMPARISON
SAE 8622 STEEL
CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE INGOT	SAMPLE CODE STRAND	STRESS LEVEL (KSI)	CONSTANTS		SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) INGOT	B50 LIFE (CYCLES) STRAND	NUMBER OF SAMPLES	
LFX	PFX	951	1.5 mm	FORGING	SIGNIFICANT DIFFERENCE AT 25% RISK	94 400	75 900	5	5
L	P	951	2.3 mm	140 mm RCS OD	SIGNIFICANT DIFFERENCE AT 10% RISK	66 200	28 800	4	5
L	P	1255	2.3 mm	140 mm RCS OD	NO SIGNIFICANT DIFFERENCE AT 10% RISK	7 540	7 120	4	6
LF	PF	951	2.3 mm	FORGING	SIGNIFICANT DIFFERENCE AT 10% RISK	44 400	76 400	5	3
LF	PF	1255	2.3 mm	FORGING	NO SIGNIFICANT DIFFERENCE AT 10% RISK	6 610	8 590	4	4

TABLE 5
HALF-RADIUS & OD SAMPLE COMPARISON
ON STRAND CAST 140 mm AND 152 mm RCS
CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE HF RAD	SAMPLE CODE OD	STRESS LEVEL (MPa)	CONSTANTS		SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) HF RAD	B50 LIFE (CYCLES) OD	NUMBER OF SAMPLES	
P1	P	951	2.3 mm	8622	SIGNIFICANT DIFFERENCE AT 10% RISK	44 100	28 800	5	5
P1	P	1255	2.3 mm	8622	SIGNIFICANT DIFFERENCE AT 5% RISK	9 620	7 120	4	6
U1X	UX	951	1.5 mm	PS16	NO SIGNIFICANT DIFFERENCE AT 25% RISK	59 000	54 700	5	6
U1X	UX	1255	1.5 mm	PS16	SIGNIFICANT DIFFERENCE AT 5% RISK	4 380	12 700	4	4
U1	U	951	2.3 mm	PS16	SIGNIFICANT DIFFERENCE AT 25% RISK	48 400	80 300	4	4
U1	U	1255	2.3 mm	PS16	NO SIGNIFICANT DIFFERENCE AT 25% RISK	8 090	8 540	6	5

TABLE 6
STEEL STOCK SIZE COMPARISON
INGOT CAST PS19 STEEL
CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE 59 mm DIA OD	SAMPLE CODE 140 mm RCS OD	STRESS LEVEL (MPa)	CONSTANTS	SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) 59 mm	B50 LIFE (CYCLES) 140 mm	NUMBER OF SAMPLES	
RX	MX	951	1.5 mm	SIGNIFICANT DIFFERENCE AT 25% RISK, NONE AT 5%	>2 000 000	339 000	4	6
RX	MX	1255	1.5 mm	SIGNIFICANT DIFFERENCE AT 25% RISK	41 500	54 300	5	5
R	M	951	2.3 mm	SIGNIFICANT DIFFERENCE AT 10% RISK	>1 775 156	106 000	4	4
R	M	1255	2.3 mm	SIGNIFICANT DIFFERENCE AT 25% RISK	26 500	21 700	4	4

TABLE 7

INDIVIDUAL SAMPLE DATA
DESIGNED EXPERIMENT

TRIAL #	UNIQUE BAR #	ALLOY LEVEL	MAX CONTROL LOAD (kN)	CYCLES TO CRACK INITIATION
10	I10RI	8627	54.4	41 472
30	F30RI	94B17	54.6	27 607
7	G7A	PS-18	54.6	36 083
12	I12RI	8627	54.3	60 838
24	K24O	PS-15	54.9	NO FAILURE
24	K24A	PS-15	54.9	NO FAILURE
32	F32RI	94B17	54.3	14 134
22	K22A	PS-15	55.0	1 116
15	L15	PS-19	55.3	8 942
4	B4A	4320 HIGH	55.0	95 650
2	B2O	4320 HIGH	55.2	550
27	J27O	8615	54.9	20 337
19	C19RI	4320 LOW	54.4	35 335
5	G5RI	PS-18	54.3	NO FAILURE
5	G5A	PS-18	54.3	NO FAILURE
25	J25A	8615	55.0	575
17	C17RI	4320 LOW	54.2	NO FAILURE
17	C17A	4320 LOW	54.2	NO FAILURE
13	L13A	PS-19	55.0	1 615
18	C18O	4320 LOW	55.1	2 993
31	F31O	94B17	55.1	13 888
8	G8O	PS-18	55.1	139 455
16	L16RI	PS-19	54.3	87 763
28	J28RI	8615	54.2	18 130
23	K23RI	PS-15	54.6	47 243
3	B3RI	4320 HIGH	54.2	57 813
11	I11O	8627	55.3	847
14	L14A	PS-19	54.5	28 614
26	J26A	8615	54.3	14 979
20	C20A	4320 LOW	55.0	NO FAILURE
20	C20B	4320 LOW	55.0	NO FAILURE
21	K21A	PS-15	54.6	154
9	I9O	8627	55.0	970
6	G6A	PS-18	55.0	80
1	B1RI	4320 HIGH	54.4	580 356
1	B1A	4320 HIGH	54.2	NO FAILURE
29	F29O	94B17	55.1	6 125

TABLE 8
1.5 mm AND 2.3 mm TOTAL CASE DEPTH COMPARISON

CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE 1.5 mm	SAMPLE CODE 2.3 mm	STRESS LEVEL (MPa)	CONSTANTS			SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) 1.5 mm	B50 LIFE (CYCLES) 2.3 mm	NUMBER OF SAMPLES 1.5 mm	NUMBER OF SAMPLES 2.3 mm
LX	L	951	INGOT	140 mm OD	8622	NO SIGNIFICANT DIFFERENCE AT <5% RISK	54 000	66 200	5	4
LX	L	1255	INGOT	140 mm OD	8622	SIGNIFICANT DIFFERENCE AT 25% RISK	5 970	7 540	3	4
LFX	LF	951	INGOT	FORGING	8622	SIGNIFICANT DIFFERENCE AT 5% RISK	94 400	44 400	5	5
PFX	PF	951	STRAND	FORGING	8622	NO SIGNIFICANT DIFFERENCE AT <5% RISK	75 900	76 400	5	3
PFX	PF	1255	STRAND	FORGING	8622	SIGNIFICANT DIFFERENCE AT 10% RISK	14 100	8 590	4	4
UX	U	951	STRAND	152 mm OD	PS16	NO SIGNIFICANT DIFFERENCE AT 5% RISK	54 700	80 300	6	4
UX	U	1255	STRAND	152 mm OD	PS16	SIGNIFICANT DIFFERENCE AT 25% RISK	12 700	8 540	4	5
U1X	U1	951	STRAND	152 mm 1/2R	PS16	NO SIGNIFICANT DIFFERENCE AT 5% RISK	59 000	48 400	5	4
U1X	U1	1255	STRAND	152 mm 1/2R	PS16	NO SIGNIFICANT DIFFERENCE AT 10% RISK	4 380	8 090	4	6
UFX	UF	951	STRAND	FORGING	PS16	SIGNIFICANT DIFFERENCE AT 25% RISK	90 200	82 600	5	4
UFX	UF	1255	STRAND	FORGING	PS16	NO SIGNIFICANT DIFFERENCE AT <5% RISK	12 700	12 000	4	4
TX	T	1255	INGOT	59 mm	8627	SIGNIFICANT DIFFERENCE AT 25% RISK	13 600	10 000	4	4
B2X	B2	1255	INGOT	102 mm OD	8627	NO SIGNIFICANT DIFFERENCE AT 10% RISK	15 800	16 700	5	5
NX	N	951	INGOT	140 mm OD	94B17	NO SIGNIFICANT DIFFERENCE AT <5% RISK	110 000	107 000	4	4
NX	N	1255	INGOT	140 mm OD	94B17	NO SIGNIFICANT DIFFERENCE AT <5% RISK	24 000	19 200	3	5
NFX	NF	951	INGOT	FORGING	94B17	SIGNIFICANT DIFFERENCE AT 25% RISK, NONE AT 5%	>2 000 000	355 000	5	5
NFX	NF	1255	INGOT	FORGING	94B17	NO SIGNIFICANT DIFFERENCE AT 5% RISK	31 800	23 800	5	5
SX	S	951	INGOT	59 mm	94B17	NO SIGNIFICANT DIFFERENCE AT 10% RISK	>1 624 605	>2 000 000	5	5
SX	S	1255	INGOT	59 mm	94B17	NO SIGNIFICANT DIFFERENCE AT 5% RISK	30 400	40 000	4	5
RX	R	951	INGOT	59 mm	PS19	NO SIGNIFICANT DIFFERENCE AT <5% RISK	>2 000 000	>1 775 156	4	4
RX	R	1255	INGOT	59 mm	PS19	SIGNIFICANT DIFFERENCE AT 10% RISK	41 500	26 500	5	4
MX	M	951	INGOT	140 mm OD	PS19	NO SIGNIFICANT DIFFERENCE AT 5% RISK	339 000	106 000	6	4
MX	M	1255	INGOT	140 mm OD	PS19	SIGNIFICANT DIFFERENCE AT 5% RISK	54 300	21 700	5	4
MFX	MF	951	INGOT	FORGING	PS19	NO SIGNIFICANT DIFFERENCE AT 10% RISK	266 000	314 000	5	5

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TABLE 9
SAE STEEL GRADE COMPARISON
CYCLE LIFE TO CRACK INITIATION

SAMPLE CODE	STEEL GRADE	SAMPLE CODE	STEEL GRADE	STRESS LEVEL (MPa)	CONSTANTS	SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) GRADE 1	B50 LIFE (CYCLES) GRADE 2	NUMBER OF SAMPLES	
LX	8622	B2X	8627	1255	INGOT 1.5 mm 102-140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	5 970	15 800	3	5
L	8622	B2	8627	1255	INGOT 2.3 mm 102-140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	7 540	16 700	4	5
P	8622	U	PS16	951	STRAND 2.3 mm 140-152 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	28 800	80 300	5	4
P	8622	U	PS16	1255	STRAND 2.3 mm 140-152 mm OD	NO SIGNIFICANT DIFFERENCE AT 10% RISK	7 120	8 540	6	5
P1	8622	U1	PS16	951	STRAND 2.3 mm 140-152 mm 1/2R	NO SIGNIFICANT DIFFERENCE AT 5% RISK	44 100	48 400	5	4
P1	8622	U1	PS16	1255	STRAND 2.3 mm 140-152 mm 1/2R	NO SIGNIFICANT DIFFERENCE AT 10% RISK	9 620	8 090	4	6
PFX	8622	UFX	PS16	951	STRAND 1.5 mm FORGING	NO SIGNIFICANT DIFFERENCE AT 10% RISK	75 900	90 200	5	5
PFX	8622	UFX	PS16	1255	STRAND 1.5 mm FORGING	NO SIGNIFICANT DIFFERENCE AT 5% RISK	14 100	12 700	4	4
PF	8622	UF	PS16	951	STRAND 2.3 mm FORGING	NO SIGNIFICANT DIFFERENCE AT <5% RISK	76 400	82 600	3	4
PF	8622	UF	PS16	1255	STRAND 2.3 mm FORGING	SIGNIFICANT DIFFERENCE AT 10% RISK	8 590	12 000	4	4
LX	8622	MX	PS19	951	INGOT 1.5 mm 140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	54 000	339 000	5	6
LX	8622	MX	PS19	1255	INGOT 1.5 mm 140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	5 970	54 300	3	5
L	8622	M	PS19	951	INGOT 2.3 mm 140 mm OD	SIGNIFICANT DIFFERENCE AT 25% RISK	66 200	106 000	4	4
L	8622	M	PS19	1255	INGOT 2.3 mm 140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	7 540	21 700	4	4
LFX	8622	MFX	PS19	951	INGOT 1.5 mm FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	94 400	266 000	5	5
LF	8622	MF	PS19	951	INGOT 2.3 mm FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	44 400	314 000	5	5
LF	8622	MF	PS19	1255	INGOT 2.3 mm FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	6 610	25 500	4	4

TABLE 9 (CONTINUED)
SAE STEEL GRADE COMPARISON
CYCLE LIFE TO CRACK INITIATION

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SAMPLE CODE	STEEL GRADE	SAMPLE CODE	STEEL GRADE	STRESS LEVEL (MPa)	CONSTANTS			SAMPLE POPULATION STATISTICAL COMPARISON	B50 LIFE (CYCLES) GRADE 1	B50 LIFE (CYCLES) GRADE 2	NUMBER OF SAMPLES	
LX	8622	NX	94B17	951	INGOT	1.5 mm	140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	54 000	110 000	5	4
LX	8622	NX	94B17	1255	INGOT	1.5 mm	140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	5 970	24 000	3	3
L	8622	N	94B17	951	INGOT	2.3 mm	140 mm OD	NO SIGNIFICANT DIFFERENCE AT <5% RISK	66 200	107 000	4	4
L	8622	N	94B17	1255	INGOT	2.3 mm	140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	7 540	19 200	4	5
LFX	8622	NFX	94B17	951	INGOT	1.5 mm	FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	94 400	>2 000 000	5	5
LF	8622	NF	94B17	951	INGOT	2.3 mm	FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	44 400	355 000	5	5
LF	8622	NF	94B17	1255	INGOT	2.3 mm	FORGING	SIGNIFICANT DIFFERENCE AT 5% RISK	6 610	23 800	4	5
TX	8627	RX	PS19	951	INGOT	1.5 mm	59 mm	SIGNIFICANT DIFFERENCE AT 10% RISK	50 500	>2 000 000	4	4
TX	8627	RX	PS19	1255	INGOT	1.5 mm	59 mm	SIGNIFICANT DIFFERENCE AT 5% RISK	13 600	41 500	4	5
T	8627	R	PS19	1255	INGOT	2.3 mm	59 mm	SIGNIFICANT DIFFERENCE AT 5% RISK	10 000	26 500	4	4
B2X	8627	MX	PS19	1255	INGOT	1.5 mm	102-140 mm OD	SIGNIFICANT DIFFERENCE AT 5% RISK	15 800	54 300	5	5
B2	8627	M	PS19	1255	INGOT	2.3 mm	102-140 mm OD	SIGNIFICANT DIFFERENCE AT 25% RISK	16 700	21 700	5	4
TX	8627	SX	94B17	951	INGOT	1.5 mm	59 mm	SIGNIFICANT DIFFERENCE AT 10% RISK	50 500	>1 624 605	4	5
TX	8627	SX	94B17	1255	INGOT	1.5 mm	59 mm	SIGNIFICANT DIFFERENCE AT 5% RISK	13 600	30 400	4	4
T	8627	S	94B17	1255	INGOT	2.3 mm	59 mm	SIGNIFICANT DIFFERENCE AT 5% RISK	10 000	40000	4	5
RX	PS19	SX	94B17	951	INGOT	1.5 mm	59 mm	NO SIGNIFICANT DIFFERENCE AT 5% RISK	>2 000 000	>1 624 605	4	5
RX	PS19	SX	94B17	1255	INGOT	1.5 mm	59 mm	SIGNIFICANT DIFFERENCE AT 10% RISK	41 500	30 400	5	4
R	PS19	S	94B17	951	INGOT	2.3 mm	59 mm	NO SIGNIFICANT DIFFERENCE AT 5% RISK	>1 775 156	>2 000 000	4	5
R	PS19	S	94B17	1255	INGOT	2.3 mm	59 mm	NO SIGNIFICANT DIFFERENCE AT <5% RISK	26 500	40 000	4	5
MFX	PS19	NFX	94B17	951	INGOT	1.5 mm	FORGING	SIGNIFICANT DIFFERENCE AT 10% RISK	266 000	>2 000 000	5	5
MF	PS19	NF	94B17	951	INGOT	2.3 mm	FORGING	NO SIGNIFICANT DIFFERENCE AT 10% RISK	314 000	355 000	5	5
MF	PS19	NF	94B17	1255	INGOT	2.3 mm	FORGING	NO SIGNIFICANT DIFFERENCE AT 10% RISK	25 500	23 800	4	5