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Influence of Phosphorus Content and Quenching/Tempering Temperatures on Fracture Toughness and Fatigue Life of SAE 5160 Steel

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This study investigates the influence of quenching/tempering temperatures on the fracture toughness and fatigue life of SAE 5160 steel, considering different phosphorus contents. Quenching and tempering treatments were applied to samples removed from different bars of commercial SAE 5160 steel with different P content. Three different austenitizing temperatures for quenching: 850, 900 and 1000 °C and a constant holding time of 15 minutes were used. The oil temperature for quenching was kept at 66 °C and the tempering conditions were 470, 500 and 530 °C with the necessary time for a final hardness of 45 ± 3 HRC. Therefore, the heat treatments cycles were applied to specimens containing low (0.012 wt. (%)), medium (0.017 wt. (%)) and high (above 0.025 wt. (%)) phosphorus contents, in order to observe the effects of this element on the susceptibility of these steels to enhance quench and tempering embrittlements. The Charpy tests results showed that the phosphorus content analyzed in this work has caused embrittlement, even in the bars with the lowest P content, leading to intergranular fracture. However, if the nucleation life is taken into consideration, this embrittlement has no effect on the nucleation fatigue life of the component.

Keywords: *5160 steel, phosphorus, austenitizing, tempering*

1. Introduction

It was observed that in the SAE 5160 steel steels used to make springs, the P content may ranges from 0.010 to 0.025 wt. (%). Manufacturers consider that this concentration does not affect the impact and the fatigue performance of this component, which is highly requested in the automotive industry[1]. By knowing the negative effects of P, this study aims to bring more light in the influence of P content, considering the regular alloying limits content found in the SAE 5160 steel bars, on both: the impact toughness and on the fatigue life.

It has been known that low-carbon alloy and plain-carbon steels are susceptible to intergranular fracture along the prior austenite grain boundaries, depending on their chemical compositions and heat treatment. In 1995 Krauss found a ductile fracture transition for intergranular fracture for steels with more than 0.5 wt. (%) carbon in low-temperature tempered martensite steels[2]. Steels with higher C content submitted to tensile or bending stresses are highly susceptible to intergranular embrittlement phenomenon under as-quenched and low-temperature tempered conditions. As tempering does not promote intergranular fracture, i.e., this embrittlement phenomenon can occur during the austenitizing or quenching process, it is called quench embrittlement (QE)[3]. The segregation of atoms of P to the prior austenite grain boundaries (a factor already known in several processes of intergranular embrittlement) and the formation of cementite (Fe_3C) around prior austenite grains are noticeable characteristics of QE[4].

The literature has a tendency towards the basic supposition that the impact toughness of hardened steels increases as the tempering temperature increases, while, at the same time, the martensite decomposition increases gradually. Thus, the minimum toughness observed may be referred to tempered martensite embrittlement (TME), which may also be identified as "irreversible embrittlement"[5], "one-step tempering embrittlement"[2] or even as "embrittlement at 350 °C"[6]. In the case of high-strength steels, TME is characterized by a significant drop in the impact toughness values in temperatures between 200-370 °C[2]. Researchers have showed that as larger is the drop in impact toughness value (V-notch Charpy impact) in this interval of temperature, higher is the TME embrittlement severity. TME occurs in a tempering interval where the ε-carbides turn into cementite (Fe_3C) and it mainly takes place in tempered martensitic microstructure. However, it is known that the same is likely to occur in steels with a bainite microstructure[6].

Another type of embrittlement is the tempering embrittlement (TE), which is a metallurgical problem found in alloy steels containing certain impurities, such as phosphorus (P), tin (Sn), antimony (Sb), and arsenic (As)[7]. These elements cause embrittlement in the mentioned steel, tempered in the temperature range of 350-570 °C, or during the low cooling rate during quenching in this temperature range. The embrittlement degree of steels depends on the impurity degree and the time in which it is kept in this temperature range. The impurities migrate to the grain boundaries during tempering period in the TE range. Relatively small amounts of these elements, as lower as 100 ppm or less have been reported as the cause for TE[2]. Other elements are also cited as embrittling agents of TE: silicon (Si), germanium (Ge), selenium (Se), tellurium (Te) and bismuth (Bi). Manganese (Mn) content may cause both, TME and TE, and

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only carbon steels containing less than 0.3% Mn are not considered susceptible to these embrittlements. Ni-Cr steels are prone to TE, but the addition of molybdenum (Mo) in amounts of until 0.5% reduces its susceptibility to TE, therefore Mo becomes an important alloy element added to steels to minimize the chances of TE occurrence[2]. The TE phenomenon can also be named "reversible embrittlement" and "two-step embrittlement"[6].

The map in Figure 1 defines the regions of intergranular and transgranular fractures, according to C and P contents for martensitic steels tempered at low temperatures, showing that the C content necessary to cause intergranular fracture, decreases as the P content increases[8]. The effect of the P content, has been shown as a result of some research that prove that the P diminish the C solubility in the previous austenite, promoting in this way, the cementite formation at the prior austenite grain boundaries.

The fact that QE being uniformly developed in steels, even if they present low P contents (lower than 0.010 wt. (%)), the rule states that the carbides formation in the grains boundaries carbides during QE, is intrinsic to this phenomenon[9].

Hyde[10] showed that the embrittlement of quenched and tempered steels may lead to a reduction in the fatigue and endurance limits due to the segregation of P and/or cementite formation in the prior austenite grains boundaries. It has been observed that in superficially hardened steels susceptible to intergranular fracture, which is the case of carburized steels, the flexural fatigue may lead to crack nucleation, with size varying from two to four grains in depth, which grows steadily in an elliptical way until the critical size be reached and the final fracture occurs instantaneously. In these cases, the nucleation was correlated to the cementite formation at the prior austenite grains boundaries, leading to a reduction in the fatigue life by reducing the nucleation stage. Figure 2 illustrate this phenomenon, where a small intergranular fracture zone occurred during the nucleation of a fatigue crack. By analyzing two cemented steels, whose carbon content in the surface was about 1 wt. (%) and P contents were 0.017 and 0.031 wt. (%). Hyde observed that the fatigue strength had been changed, so the embrittlement affected more significantly the fatigue limit (Figure 3).

The quenching of high-strength steels to obtain harder martensitic structures is generally expected for high cycle fatigue regime and high wear performance. Nevertheless, this process can introduce microcracking and/or promote intergranular cracking nucleation at the grain boundaries[10-12]. In a brittle material, the strength of the grains is usually weaker than its interface, therefore cracks nucleate in transgranular way. Even so, with increasing hardening, the strength in the interior of a grain, usually becomes higher than at the interface (boundary), nucleating cracks intergranularly. Corresponding to this mechanism is the intergranular cracking by fatigue in low-carbon martensitic steels, where the proeutectoid ferrite precipitates at the prior austenite grains boundaries, weakening them[13]. In the case of carburized high-resistance steels, a flaw may occur intergranularly at the prior austenite grains boundaries, depending on the severity of the embrittlement[14].

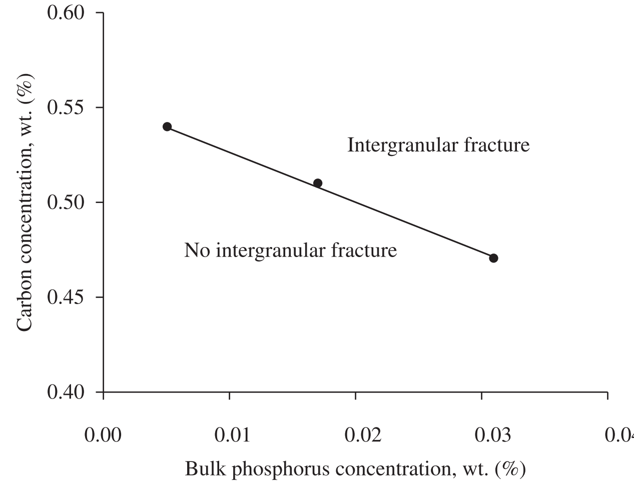


Figure 1. Fracture mode due to P and C contents for martensite tempered at low temperatures[8].

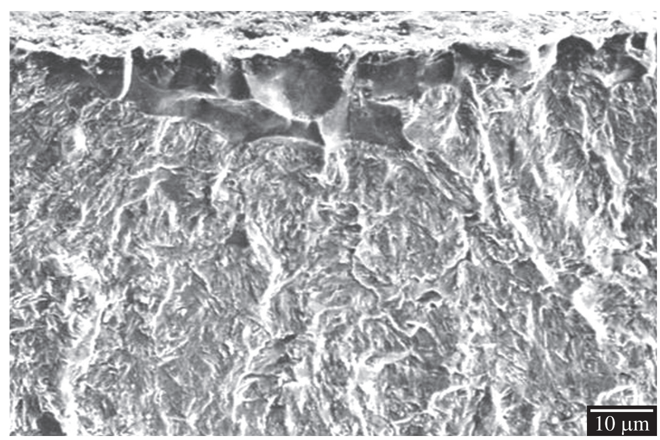


Figure 2. Location of the intergranular initiation of a fatigue fracture surface of a specimen of SAE 4320 carburized steel[10].

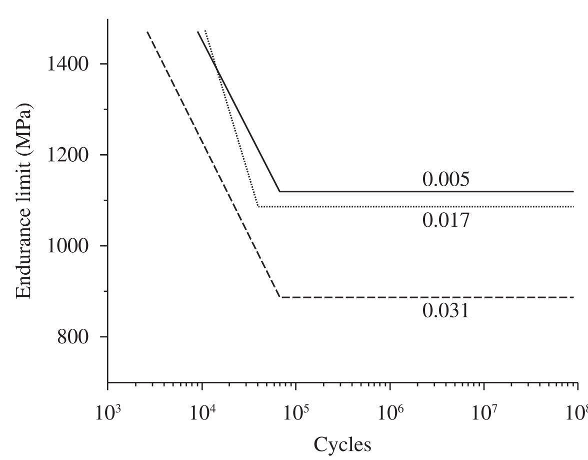


Figure 3. Comparison of maximum stress versus number of cycles for steels with different phosphorus contents[10].

2. Experimental

2.1.Material

The material investigated in the present work was a SAE 5160 steel, with four different P content and provided as laminated bars with 70 × 10 mm^2 thickness and 1000 mm length. Its chemical composition is presented in Table 1.

2.2. Heat treatments

Figure 4 presents a schematic flowchart indicating the heating treatments performed in the SAE 5160 steel bars, from which specimens were removed for hardness, impact toughness and fatigue (S-N) tests. These bars were gradually heated in a continuous two-heating zone furnace. The soaking time at the second zone was 15 minutes. All bars, after austenitizing, were quenched in mineral oil and tempered. Basically, three different austenitizing temperatures and three different tempering temperatures were studied for the four different levels of phosphorus studied. The quenching oil was always kept at 66 °C.

2.3. Mechanical tests

As mentioned before, the hardness, impact, fracture toughness and fatigue (S-N) tests were performed to investigate the susceptibility of the SAE 5160, with different allowable P content, to the embrittlement effect, identifying the more adequate process parameters, as quenching and tempering temperatures, to avoid these embrittlements.

Rockwell C hardness tests were conducted in a Leco RT-240 150 kgf durometer using a 10 mm diameter conic penetrator (120°) made from diamond, according to ASTM E18-04. The hardness values represent an average of 10 measurements, acquired in the width of the surface in each sample.

For the impact tests, type A Charpy samples with V-notch were machined according to ASTM E23-04, as shown in Figure 5. The fabrication was performed in a milling machine and the V-notch in a special backed-off cutter machine. The test specimens were removed in the L-S direction, so that the notch was parallel to the width of the sheet steel. This is the preferred direction for the in-service crack propagation. The tests were performed in an Instron Wolpert PW30 impact test device, using a 300 J maximum energy capacity and impact speed of 5.54 m/s.

The fatigue tests were performed in a 1 ton nominal capacity Guttman Eccentric Press (used for fatigue tests of flat springs). This machine uses a three-point bend device in which the specimen is placed over two points separated by 650 mm span, as shown in Figure 6. The maximum load applied was 1093 kN (corresponding to a maximum tension stress of $\sigma_{máx}$ = 130 MPa in the center of the bar at surface), which corresponds to a load necessary to cause a 25 mm deflection in the center of the bar. It is important to mention that all specimens, underwent to a shot-peening process on the surface that during test would be under tensile stresses, before they were fatigue-tested, similarly to the procedure for regular leaf springs.

The three points fracture toughness tests were performed according to ASTM E1820-06. The load ratio, maximum load and frequency used in the fatigue pre-cracking of specimens were, R = 0.1, $P_{máx}$ = 5.5 kN and f = 30 Hz, respectively.

A servo-hydraulic MTS810 system was used to conduct the fracture toughness test. As the bar thickness was only 10 mm, the specimens was removed in the T-L direction and three-point bend specimens, SE(B) were used with B = 10 mm; W = 20 mm; a_0 = 9.5; S_d = 80 mm; and f(a/W) = 2.46, as shown in Figure 7. The rate of applied load was 13 kN/min and the crack opening displacement was measured using an extensometer fixed at the crack mouth.

Initially, two alloys were previously chosen in an attempt to validate the obtained K_{IC} values. The criterion was the selection of specimens heat treated for the more fragile and more ductile conditions. For validation purpose, it is worthwhile to emphasize that the blade gage restricted the specimen to its thickness (B) = 10 mm.

As stated by the ASTM 1820-06 standard, the condition that must be satisfied is:

$$a, B, (W-a) \geq 2,5\left(\frac{K_{IC}}{\sigma_e}\right)^2 \quad (1)$$

2.4. Grain size analysis

Samples removed from the bars were prepared according to ASTM E3-08. After the surface preparation, a chemical attack

Table 1. Chemical composition of SAE 5160 steel wt. (%).

	Specification SAE	Steel A	Steel B	Steel C	Steel D
C	0.56 – 0.64	0.593	0.580	0.584	0.600
Mn	0.75 – 1.00	0.860	0.850	0.850	0.860
P	0.030 (máx)	0.012	0.017	0.025	0.031
S	0.035 (máx)	0.012	0.011	0.008	0.005
Si	0.15 – 0.35	0.250	0.230	0.012	0.190
Cr	0.70 – 0.90	0.800	0.770	0.780	0.770
Ni	0.00 – 0.25	0.070	0.090	0.070	0.090

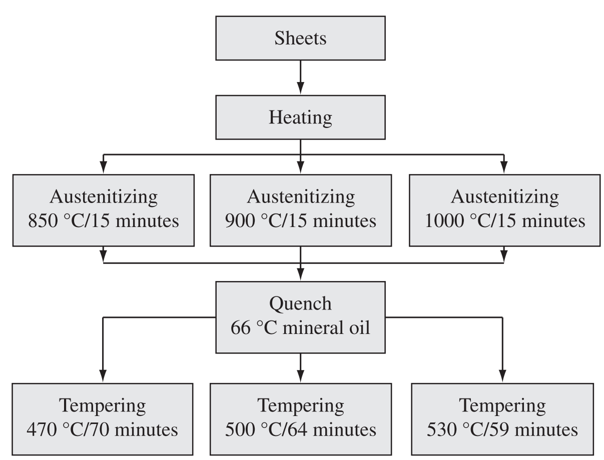


Figure 4. Schematic flowchart of the heating treatment routs for the SAE 5160 steel bars.

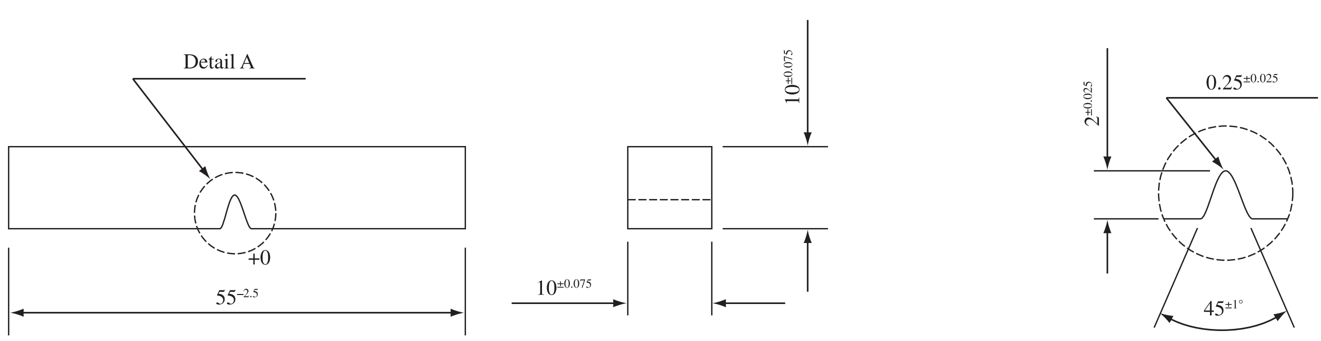


Figure 5. Size and geometry of the type A Charpy specimens according to ASTM E23-04.

was performed on the surface of each sample. The chemical attack consisted in a solution of 1.0 g of picric acid, 1.0 mL of chloride acid, 15.0 mL of detergent + 100.0 mL of distilled water, over a time of six minutes. Afterwards, each sample was observed in a Zeiss Axiotech Stemi DV4/DR microscope connected to Graftek Ultimage Optlab software, Version 2.6. The grain size was measured according to ASTM E-1382-04.

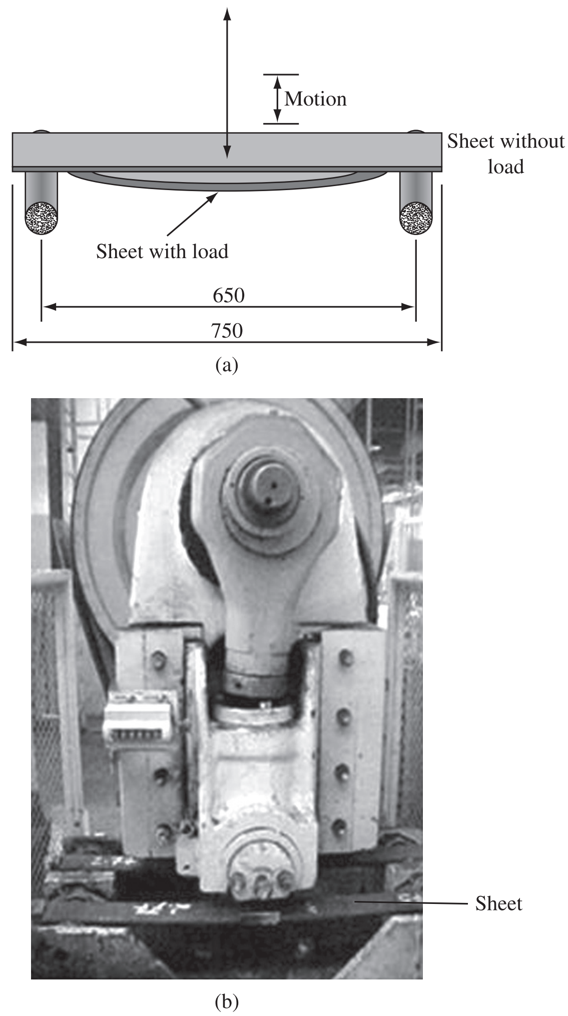


Figure 6. Schematic representation of the fatigue test. The dimensions are in mm.

2.5. Fractographic analysis

The fracture surface of the broken specimens after impact, fracture toughness and fatigue tests, were observed in a DSM 960 ZE155 scanning electronic microscope. The fracture surfaces have been previously cleaned with acetone in an ultrasonic equipment, so that no undesirable impurity would influence the analysis. This type of analysis is required to evaluate the micro-mechanism of fractures.

3. Results and Discussion

As observed in Table 1, except for the P content, the chemical compositions of the four steels are alike. The P content was below the maximum allowable value always.

Table 2 shows the values of grain sizes for each condition. The invariability of the grain size was expected since these steels were aluminum killed to avoid large grains.

Figure 8 shows the mean hardness values obtained from each steel composition and heat treatment as presented in the flow chart in Figure 4. For the same austenitizing temperature, the hardness values decrease for an increasing tempering temperature, However, it is seen that the dependence on tempering temperature is reduced as the austenitizing temperature for quenched increased from 850 to 1000 °C. Most hardness values of the tested samples are inside the limits specified for the leaf springs produced from this steel, i.e., 43-48 HRC. For the steel D (0.031% of P) tempered at 530 °C, the hardness value was slightly below the minimum value (43 HRC). These lower hardness values produced higher impact energy values than those expected for the SAE heat treated for the nominal hardness value (43-48 HRC). Unfortunately, at the highest tempering temperature (530 °C), the furnace did not allowed further reduction in the time range (due to speed of stretching of the oven) to provide the necessary hardness value.

Figure 9 shows an overall representation of the absorbed impact energy for all heat treatment conditions, considering the different levels of P investigated. It is observed that, independent of the austenitizing temperature for quenching, lower the P content, higher were the values of absorbed impact energy, especially for the tempering temperature of 500 °C. Also, steel A (0.012 P) austenitized at 850 °C and tempered at 500 °C exhibited the highest value of absorbed impact energy, approximately 400 $kJ.m^{-2}$, despite its highest

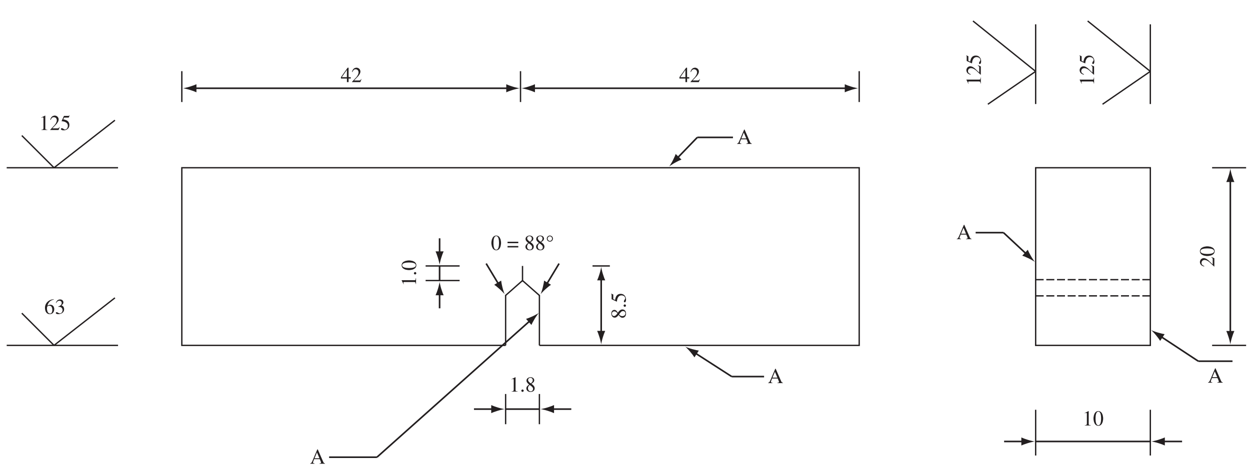


Figure 7. SE (B) specimen dimensions and geometry used for fracture toughness tests. Dimensions are in mm.

Table 2. Grain size ASTM (μm) for each studied condition.

Steel A (0.012% P)		$T_{tempering}$		
		470 °C	500 °C	530 °C
$T_{quenching}$	850 °C	---	24.4 ± 0.1	---
	900 °C	---	25.0 ± 0.3	---
	1000 °C	---	23.9 ± 0.3	---
Steel B (0.017% P)		$T_{tempering}$		
		470 °C	500 °C	530 °C
$T_{quenching}$	850 °C	24.9 ± 0.3	24.3 ± 0.4	25.0 ± 0.1
	900 °C	24.1 ± 0.2	23.5 ± 0.1	24.0 ± 0.1
	1000 °C	23.0 ± 0.1	22.9 ± 0.1	22.8 ± 0.3
Steel C (0.025% P)		$T_{tempering}$		
		470 °C	500 °C	530 °C
$T_{quenching}$	850 °C	---	24.9 ± 0.3	---
	900 °C	---	24.4 ± 0.2	---
	1000 °C	---	23.8 ± 0.2	---
Steel D (0.031% P)		$T_{tempering}$		
		470 °C	500 °C	530 °C
$T_{quenching}$	850 °C	24.7 ± 0.1	24.2 ± 0.2	23.6 ± 0.1
	900 °C	25.0 ± 0.1	23.7 ± 0.2	24.1 ± 0.2
	1000 °C	22.8 ± 0.2	23.2 ± 0.2	22.5 ± 0.2

Obs: --- means that no measurement was made.

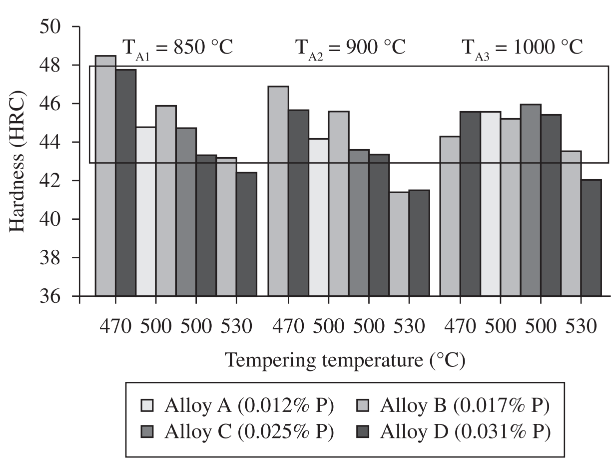


Figure 8. Hardness test results versus tempering temperature for three austenitizing temperatures (TA_1, TA_2, and TA_3).

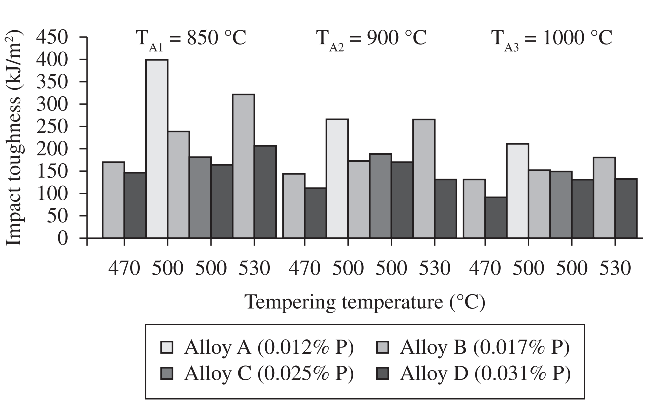


Figure 9. Absorbed energy in the Charpy impact test. Comparative image for all conditions studied in this work.

hardness values when compared to the ones obtained at austenitizing temperature of 900 °C.

According to Reguly[4], for the SAE 5160 steel an increase in the tempering temperature, will cause an improvement on the impact toughness, if the phosphorus content is low (0.001 wt. (%)), since it will not develop quench embrittlement. In the case that the P content was low, but developed quench embrittlement after austenitizing at 1100 °C, there was an improvement in the impact toughness when the tempering temperature was high. Reguly[4], observed also that improvement in the impact toughness values were obtained when the austenitizing temperature was below 1100 °C.

As mentioned before, the hardness values of any specimen tempered at 530 °C were not in the required hardness interval (43-48 HRC), as a consequence the toughness values were overestimated and the fracture surfaces of these specimens were not considered for the analyses in this article.

Figure 10 shows the fractographic images of the Charpy impact specimens tempered at 470 °C for 70 minutes. For Steel D (0.031 wt. (%) P) the intergranular fracture prevails at all austenitizing temperatures, while the intergranular fracture of steel B (0.017 wt. (%) P) occurs only for austenitizing at 1000 °C. The austenitizing temperatures of 850 and 900 °C for steel B (0.017 P) produced a region of mixed fracture (ductile + intergranular) with a dominance of ductile type. The amount of intergranular fracture gradually increases as the austenitizing temperature rises, explaining the gradual drop in the impact toughness values, according to the increase in the austenitizing temperature for both steels B and D for the studied steels, Figure 9.

Figures 11 and 12 show images of impact specimens tempered at 500 °C for 64 minutes. In steels A (0.012 wt. (%) P) and C (0.025 wt. (%) P). Figure 12 features a ductile fracture of steel B (0.017 wt. (%) P) and also a mixed aspect in the specimen from steel D (0.031 wt. (%) P), which shows mostly ductile with only some regions of intergranular fracture. This intergranular portion is directly connected to the 33% decrease in the impact toughness if compared to specimens from steel B (0.017 wt. (%) P), which display an absorbed impact energy of 240.54 $kJ.m^{-2}$.

Thus, for the austenitizing temperature of 850 °C it is clearly noticed that increasing the P content, the amount of intergranular fracture increased and consequently the value of the absorbed impact energy was reduced. Yet, for those cases of austenitizing temperature of 900 °C, Figure 12 shows the predominance of intergranular fracture for specimens from steel D with high P content (0.031 wt. (%)), and predominance of ductile fracture for specimens taken from steel B with medium P content (0.017 wt. (%)). However, unexpectedly the impact toughness values of both steels were quite similar, i.e., 173 and 168 $kJ.m^{-2}$ for steels B and D, respectively.

At austenitizing temperature of 1000 °C, steel B (0.017 wt. (%) P) exhibited a significantly larger portion of intergranular fracture than the specimens austenitized at 900 °C. Such fact is evidenced by the fall of approximately 13% in the value of impact toughness when the autenitizing temperature was 1000 °C. On the other hand, the sample of steel D (0.031 wt. (%) P) displayed a fracture surface with predominantly intergranular type.

In short, the fractographic analysis agrees the fact that the values of impact toughness of the steels with higher levels of phosphorus are always lower than those with lower levels of P (Figure 9). The large portion of intergranular fracture found in steels with higher levels of P is related to the decrease in the values of absorbed impact energy. It can be inferred that this embrittlement phenomenon is due to the segregation of phosphorus occurred during the respective austenitizing and/or tempering heat treatments, for steels with higher P content[9].

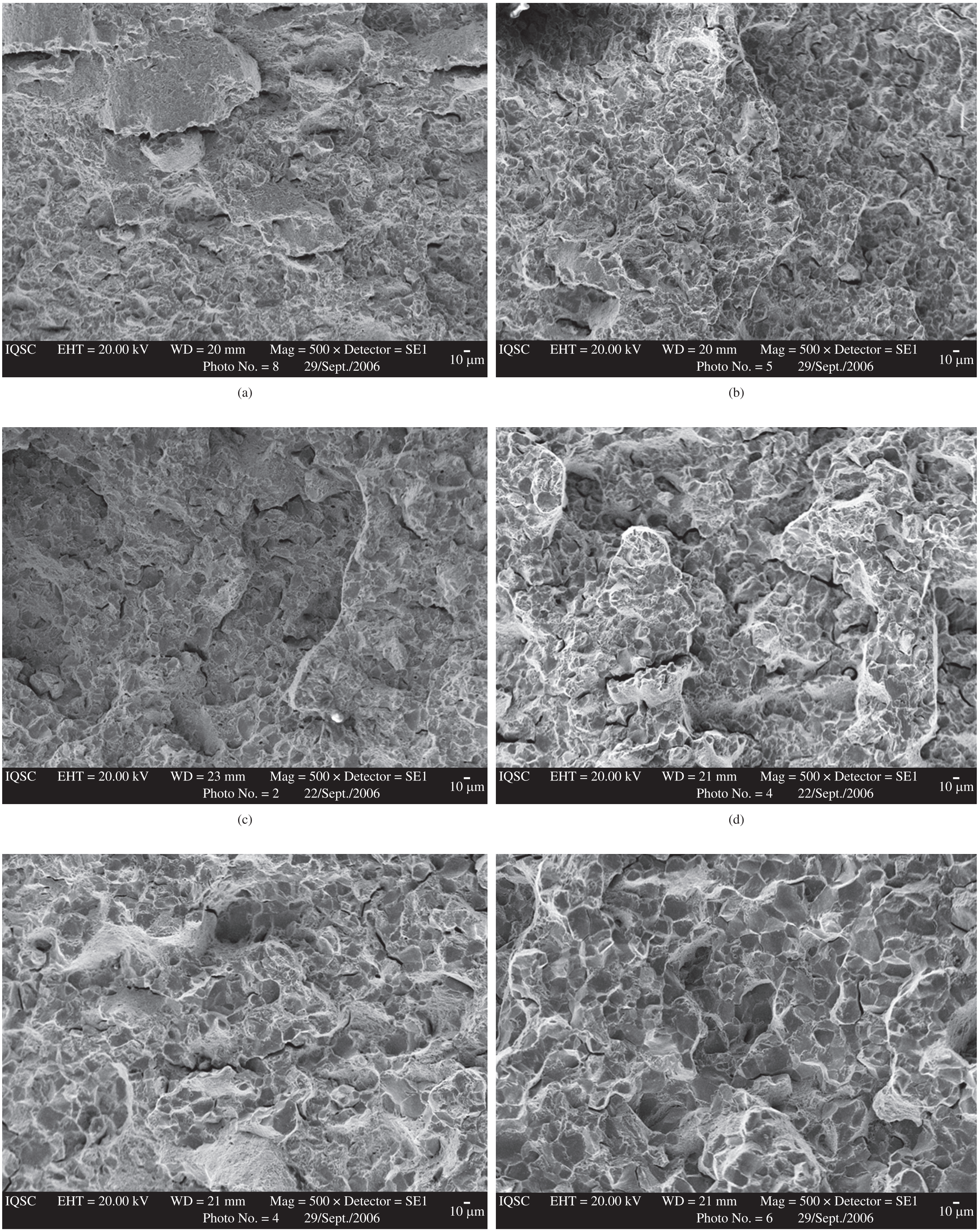


Figure 10. Fracture surface (central part of the fracture) of impact specimens. Steel with 0.017 wt. (%) P (steel B) austenitized at: a) 850 °C; c) 900 °C; and e) 1000 °C. Steel with 0.031 wt. (%) P (steel D) austenitized at: b) 850 °C; d) 900 °C; and f) 1000 °C. The specimens were tempered at 470 °C for 70 minutes.

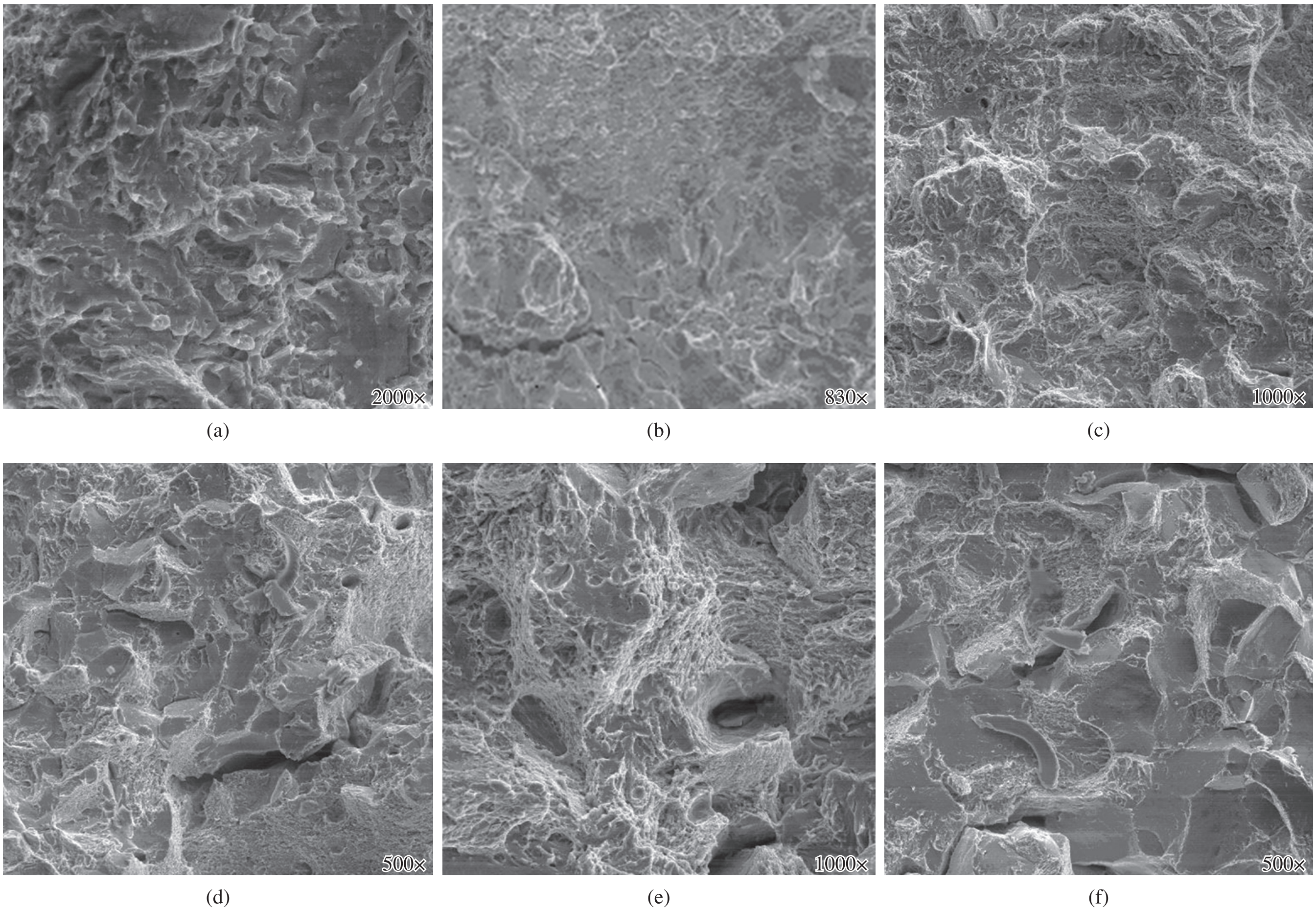


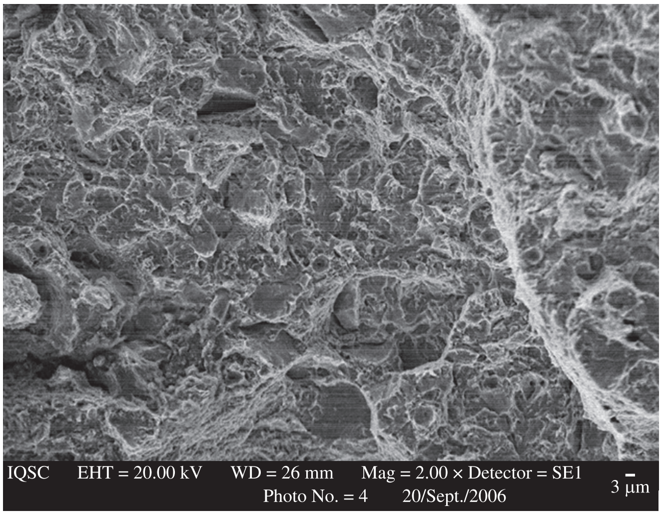
Figure 11. Fracture surface (central part of the fracture) of the impact specimens. Steel with 0.012 wt. (%) P (steel A) austenitized at: a) 850 °C; c) 900 °C; and e) 1000 °C. Steel with 0.025 wt. (%) P (steel C) austenitized at: b) 850 °C; d) 900 °C; and f) 1000 °C. The specimens were tempered at 500 °C for 64 minutes.

Figure 13a shows the Load versus COD curve for steel B (0.017 wt. (%) P) austenitized at 850 °C and tempered at 530 °C. According to ASTM E1820-2006, it is observed that the curve for steel B is from Type I, meaning that the value of P_Q is lower than the value of $P_{máx}$. Thus, it is necessary to use the procedure of the 5% secant line OP_5. Figure 13b shows the behavior of a steel D (0.031 wt. (%) of P) austenitized at 1000 °C and tempered at 470 °C. The Load versus COD curve in this case is from type III, with elastic behavior an indicative of low toughness.

Table 3 shows the fracture toughness results for specimens from steel B (0.017 wt. (%) of P) and steel D (0.031 wt. (%) of P). The last line of Table 3 shows the results of a, (W–a) and B, which are the requirements for K_Q be considered the fracture toughness value in plane strain state condition, K_{IC}. It is observed that B = 10.0 mm did not allowed the obtaining of a valid value of K_{IC} for steel B (austenitized at 850 °C and tempered at 530 °C), therefore it is considered as K_c only. However, for steel D (austenitized at 1000 °C and tempered at 470 °C), it was possible to obtain $K = K_{IC}$, therefore only the values of K_Q were used for toughness for these tested conditions. Note that, similarly to the impact test trend with P content, the values of K_Q decreases as the phosphorus level increases.

The values of K_Q relates well with the fracture surface results, i.e., steel B, which had a higher value of K_Q, presented a ductile fracture mode, while steel D, which had a lower value of K_Q, presented a brittle fracture mode. The fracture surface from steel B, Figure 14, shows a ductile fracture mode, with stable crack growth by coalescence of microvoids, Figure 14b. In this specimen it was also observed a delamination, where the internal faces presents a more brittle aspect, Figure 14c. From Figure 15 it is seen that steel D exhibited a more brittle mode of fracture, with a mixture of cleavage and intergranular types. Such a fracture micromechanism occurred due its highest P content, even considering that it is inside of the allowable range. As previously observed by other researchers[9-11], atoms of P may migrate to prior austenitic grain boundaries, making the steel to behave in a brittle manner. However, it is not possible to precise in which step it has occurred, if during austenitizing or tempering.

Figure 16 presents the results of the fatigue tests. Note that, for similar heat treatment, there is a spread in the measured values of fatigue life. This scattering could be related to the stress concentrators in the spring surface and/or local changes in mechanical properties, considering that the specimens were directly acquired from bars made on industrial scale. However, besides the austenitizing and tempering temperatures, the values of high cycle fatigue life were not significantly affected by the embrittlement caused by a higher level of P, as the heat treatments aimed to produce micro-structures able to provide similar mechanical properties, such as hardness and tensile strength to the specimens, consequently causing a similar high cycle fatigue life.



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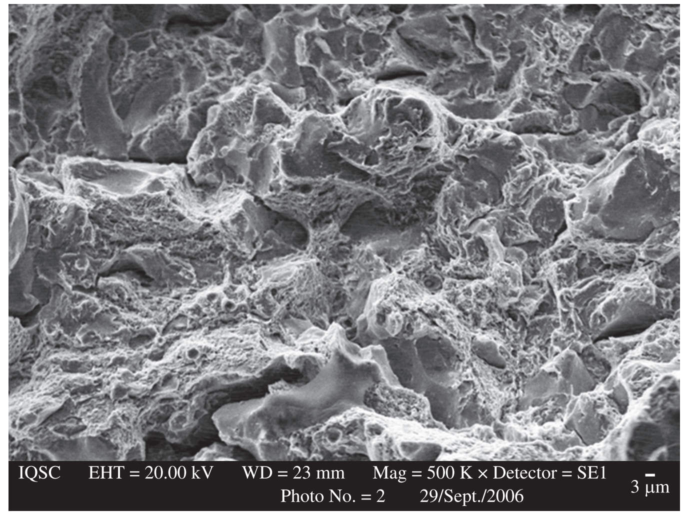

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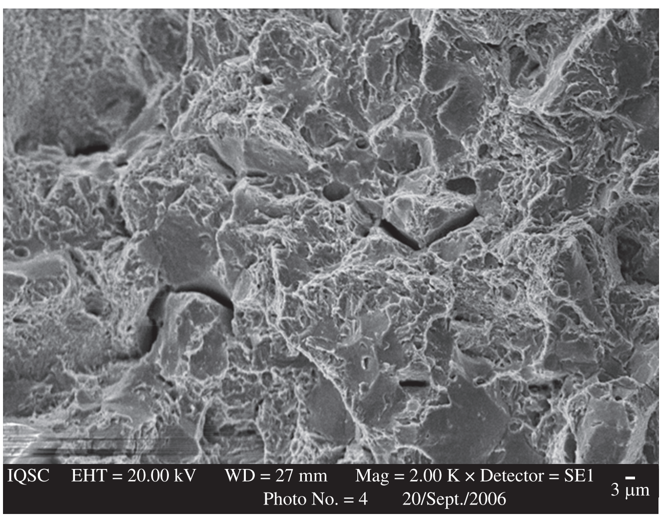

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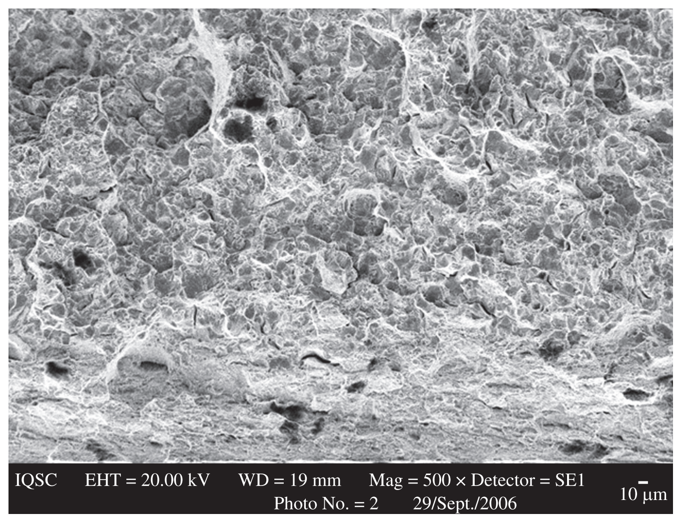

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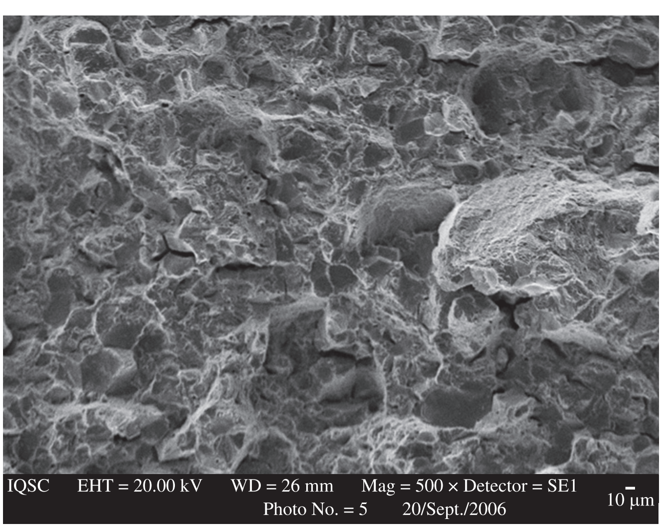
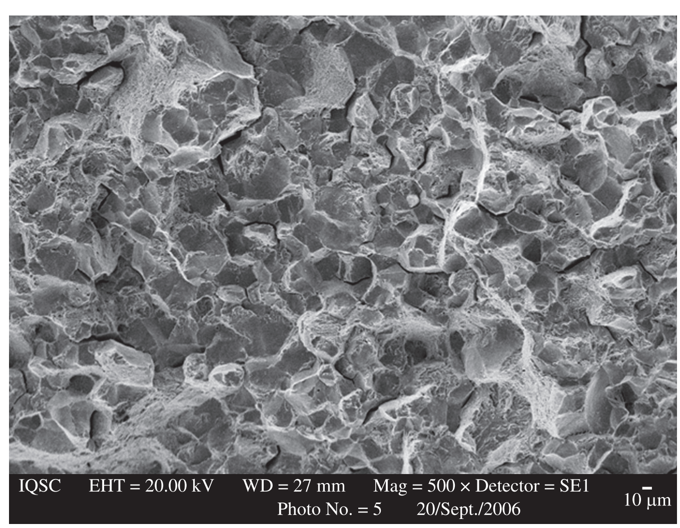


(e)



(f)

Figure 12. Fracture surface (central part of the fracture) of the impact specimens. Steels with 0.017 wt. (%) P (steel B) austenitized at: a) 850 °C; c) 900 °C; and e) 1000 °C. Steels with 0.031 wt. (%) P (steel D) austenitized at: b) 850 °C; d) 900 °C; and f) 1000 °C. The specimens were tempered at 500 °C for 64 minutes.

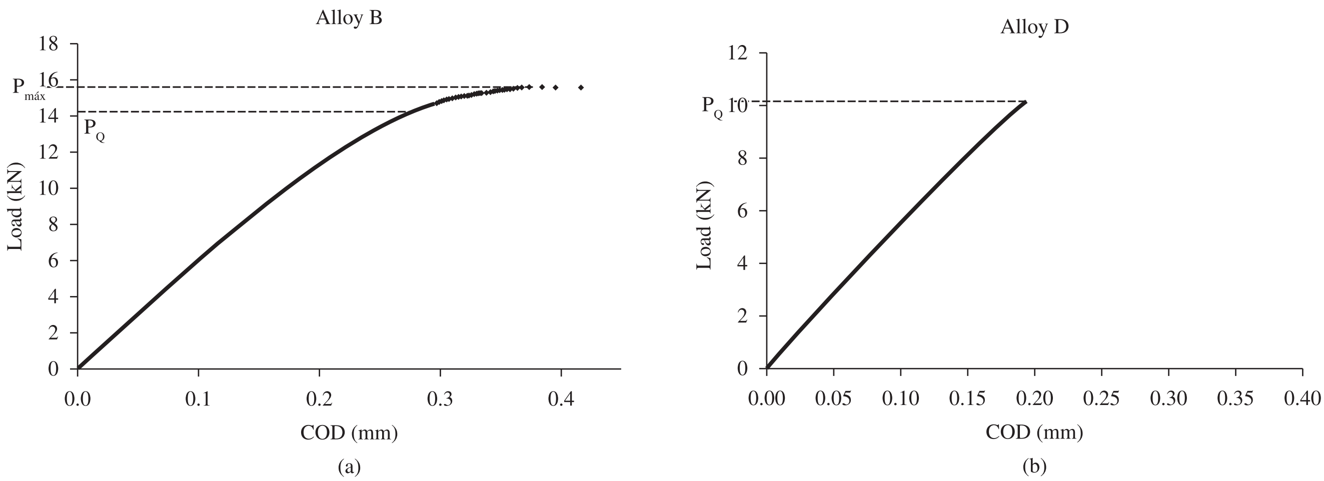


Figure 13. Load versus crack opening displacement form SEN(B) specimens for determination of KQ: a) steel B (0.017 wt. (%) of P) austenitized at 850 °C and tempered at 530 °C and b) steel D (0.031 wt. (%) of P) austenitized at 1000 °C and tempered at 470 °C.

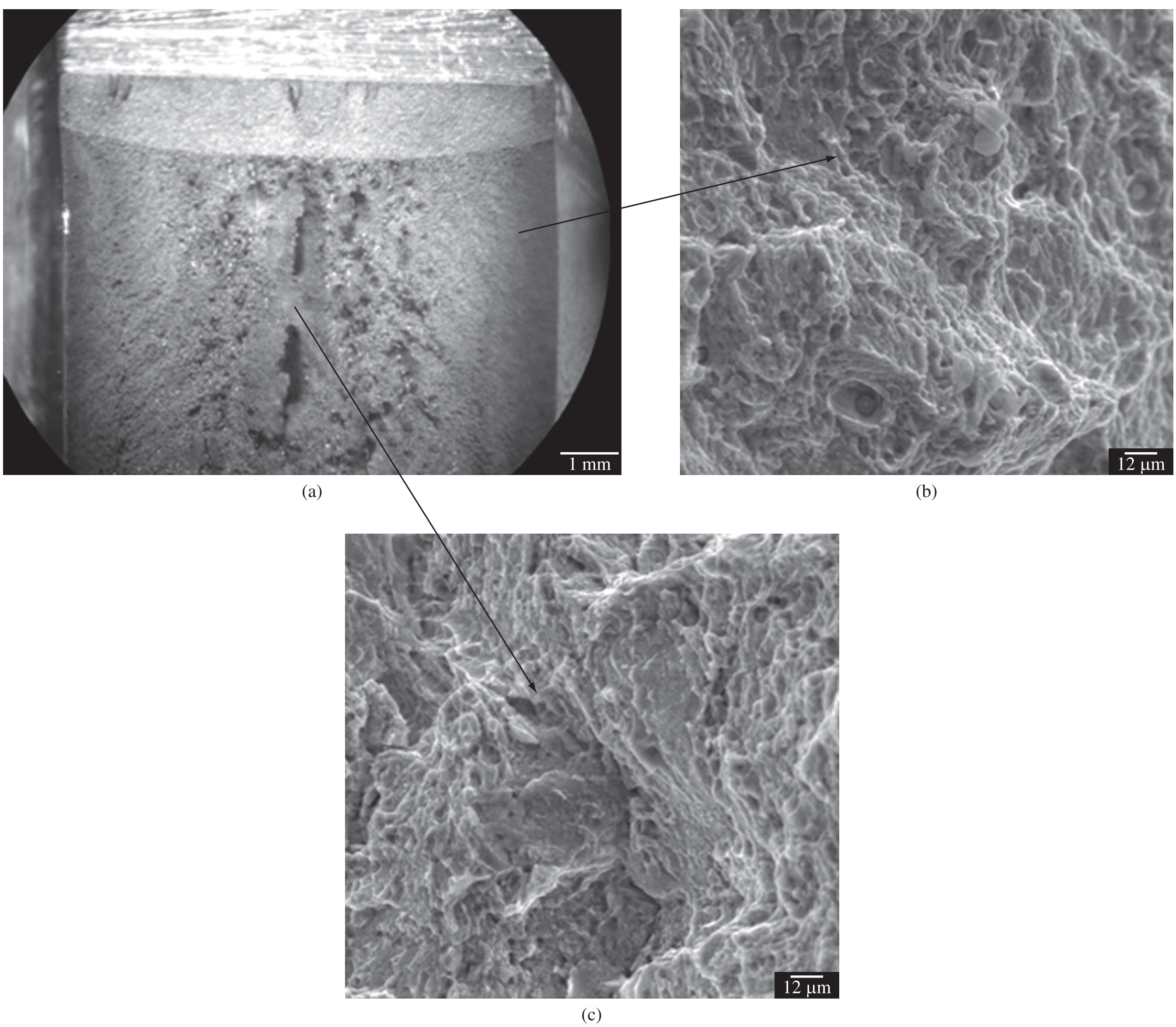
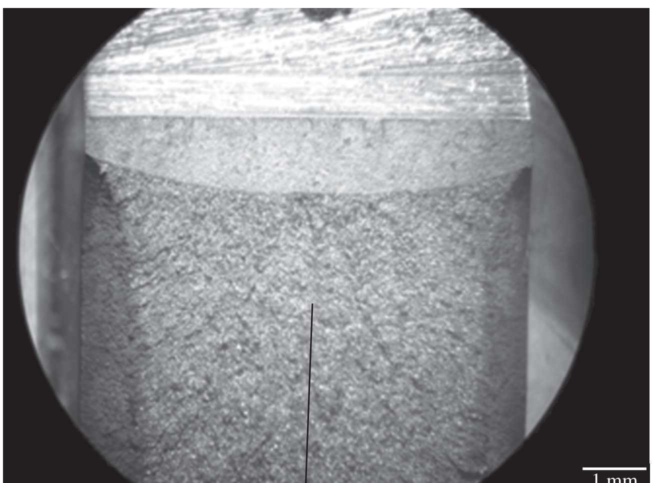


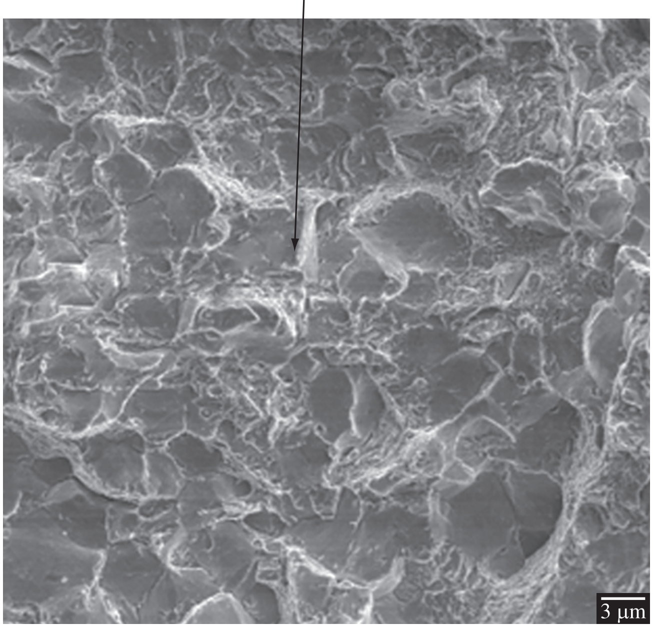
Figure 14. Fracture surface from steel B (0.017% of P) austenitized at 850 °C and tempered at 530 °C. a) General view of the fracture surface; b) detail from the instable crack growth; and c) detail from the fracture mode from the delamination.

Table 3. Mechanical properties from tensile test, austenitizing and tempering temperatures and values of load (P_Q) and toughness (K_Q) for steels B and D.

Steels	B (0.017 wt. (%) P)	D (0.031 wt. (%) P)
Quenching temperature (°C)	850	1000
Tempering temperature (°C)	530	470
Yield strength (MPa)	1262.3 ± 8.5	1430.3 ± 9.2
Ultimate strength (MPa)	1327.3 ± 15.9	1547.0 ± 6.2
P_Q (kN)	14.2 ± 0.4	10.3 ± 0.3
K_Q [$MPa\sqrt{m}$]	98.8	71.7
(W-a), B, a (mm)	15.3	6.3

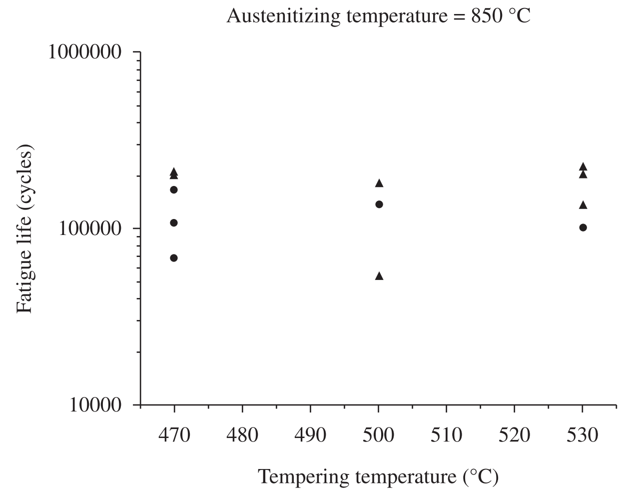


(a)

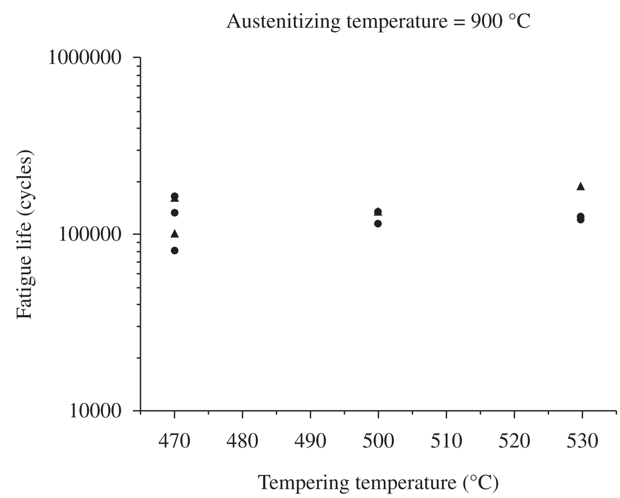


(b)

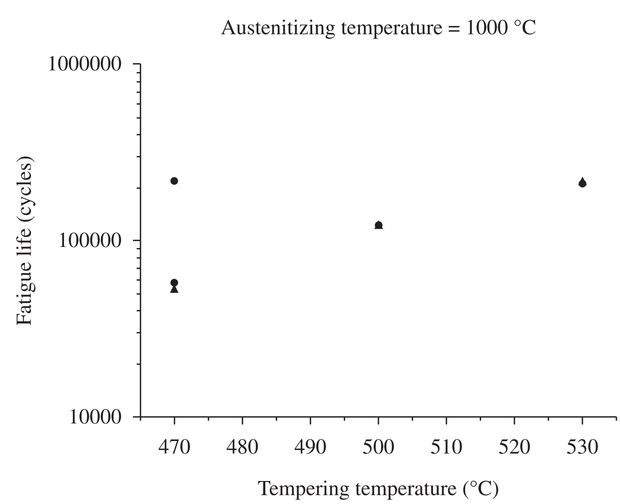
Figure 15. Fracture surface from steel D (0.031% of P) austenitized at 1000 °C and tempered at 470 °C: a) General view showing the notch, fatigue pre-crack and instable crack growth and b) Detail from the instable crack growth showing a brittle fracture mode.



(a)



(b)



(c)

▲ 0.017% P ● 0.031% P

Figure 16. Fatigue test results (S-N) performed in the heat-treated samples for austenitizing temperatures of a) 850 °C; b) 900 °C; and c) 1000 °C and a σmáx = 130 MPa applied in the center of the sheet and R = 0.1.

4. Conclusions

From the results it is possible to conclude that the P content has a strong influence on the impact properties of the SAE 5160 steel, even if considered P content below the maximum allowable value (0.03 wt. (%)). The effect is more pronounced at lower quenching temperature (850 °C). For higher quenching temperatures (900 and 1000 °C), the drop in the impact toughness is smaller.

Although, it was not possible to validate the fracture toughness results for steel B (quenched from 850 °C and tempered at 530 °C) following the ASTM E1820, the K_Q values were able to show a 30% drop in the stress intensity factor value when compared with the brittle steel (steel D quenched from 1000 °C and tempered at 470 °C).

The fractographic analysis relates quite well with the absorbed energy and stress intensity factor values obtained for the steels analyzed, i.e., higher P content was related to larger percentile of intergranular fracture, as well as with increasing temperature for quenching.

Independently of the austenitizing and tempering temperatures, the values of high cycle fatigue life were not significantly affected by the embrittlement caused by higher levels of P.

As a final conclusion, it was noticed that the P content has a very deleterious effect on the SAE 5160 steel and it is strongly recommended that spring steels must have P content as low as 0.01 wt. (%).

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