# FRACTURE TOUGHNESS OF ROLLING BEARING STEELS

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This paper describes a set of alternatives heat treatments on SAE 52100 rolling bearing steel. It is shown that shorter times and higher temperatures in austenitization and tempering allow the attainement of equivalent fracture toughness values and equal or better hardness characteristics than the reference industrial heat treatment. Fractographic examinations explain the toughness results, showing that the fracture mechanism is quite insensitive to heat treatment variables during austenitization.

## INTRODUCTION

In the case of the through-hardening process of SAE 52100 steel (UNI 100Cr6), long durations required for the austenitizing and tempering treatments call for a special plant installation off line, thus causing discontinuity in the global production worktable. This yields the need of investigations on innovative on-line heat treatments, characterized by shorter durations and higher temperatures. On the other hand, in order to guarantee a correct and reliable rolling bearing performances, adequate toughness ($K_{IC}$) has to be reached to tolerate the mounting, applied stresses regimes and, most of all, tolerance to shock loads [1]. Moreover, in certain high speed applications (e.g. mainshaft of turbines), where hertzian stresses are low but tensile hoop stresses are present, it is of particular importance that small fatigue cracks can reach a detectable size before

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failure [2]: once again it is necessary that fracture toughness be not too low. A study has been undertaken to assess the feasibility of "on line" heat treatment cycles, based on both short time high temperature austenitizing and small tempering stages [3]. Whereas the proof of the feasibility of a calibrated cost saving heat treatment has already been obtained [3], in this paper the attention is focused on the fracture aspects to obtain some insight on the correlation between heat treatments variables, material microstructure and dominant fracture mechanism.

## EXPERIMENTAL

The heat treatments were performed in salt baths, with austenitization temperature ranging from 850 to 940°C, followed by salt bath quenching at 155°C for 10' and with tempering between 150 and 250°C; austenitizing time being adjusted to obtain the same hardness and equivalent final microstructure. Previous results [3] and data reported in literature [4] point out that isothermal quenching results in a negligible retained austenite content. For the subsequent tempering, owing to the complexity of the program, the choice of the treatment parameters was simply limited to 180, 210 and 250°C, for 1, 2, 5 and 10'. Among all the possible interrelated combinations considered here for quenching and tempering, only those having the same or better hardness and toughness, compared with the classical treatment, have been realized. Austenitization at 850°C for 17', followed by oil quenching at 50°C and tempering at 155°C for 1 h, was also performed, thus obtaining the classical reference heat treatment.

Two lots of spheroidizing annealed UNI 100Cr6 steel (SAE 52100) were employed (C=1%, Mn=0.35%, Si=0.3%, Cr=1.5%), differing only in the sulphur and phosphorus content: the clean one with S=0.008% and P=0.007%, the other one, labelled (*), with S=0.013% and P=0.016%. Vickers hardness tests were carried out on heat treated rings 1 mm thick with 25 mm inner diameter and 40 mm outer diameter. The small thickness was chosen in order to minimize thermal transient during heat treatment. Fracture toughness tests were conducted on heat treated CT samples (W=25 mm, B=6 mm). In fracture toughness CT samples, time and temperature of the heat treatments were taken in the centre, owing to the presence of thermal transient. $K_{IC}$ was determined only at time-temperature combinations leading to satisfactory hardness and microstructure. Fracture surfaces were investigated by SEM.

## RESULTS

Hardness tests results referring to the austenitization treatment are reported in Figure 1 for each temperature as a function of time. As expected, hardness increases with austenitizing time because this implies increasing carbon content in the martensite. Hardess values are almost everywhere higher than the 790 HV

TABLE 1- Fracture toughness results; the medium phosphorus steel is labelled by *.

| Aust T (°C) | Aust t (s) | Cooling T (°C) | Medium | Tempering | K IC [Mpa*m^1/2] | Hardness HV |
|---|---|---|---|---|---|---|
| 940 | 1' 04" | 155 | salt 10'/air | - | 13,22 | 917 |
| 925 | 16" | 155 | salt 10'/air | - | 13,55 | 897 |
| 925 | 32" | 155 | salt 10'/air | - | 12,92 | 905 |
| 925 | 1' 04" | 155 | salt 10'/air | - | 13,36 | 910 |
| 925 | 2' 08" | 155 | salt 10'/air | - | 12,47 | 914 |
| * 925 | 2' 08" | 155 | salt 10'/air | - | 10,72 | 914 |
| 910 | 1' 04" | 155 | salt 10'/air | - | 14,02 | 898 |
| 910 | 2' 08" | 155 | salt 10'/air | - | 12,07 | 902 |
| * 910 | 2'08" | 155 | salt 10'/air | - | 11,37 | 902 |
| 910 | 4' 16" | 155 | salt 10'/air | - | 11,1 | 906 |
| * 910 | 4' 16" | 155 | salt 10'/air | - | 11,25 | 906 |
| 895 | 32" | 155 | salt 10'/air | - | 14,14 | 863 |
| 895 | 1' 04" | 155 | salt 10'/air | - | 14,38 | 881 |
| 895 | 2' 08" | 155 | salt 10'/air | - | 13,93 | 889 |
| 895 | 4' 16" | 155 | salt 10'/air | - | 11,2 | 895 |
| * 895 | 4' 16" | 155 | salt 10'/air | - | 11,4 | 895 |
| 895 | 8' 32" | 155 | salt 10'/air | - | 13,85 | 901 |
| 880 | 4' 16" | 155 | salt 10'/air | - | 13,06 | 877 |
| 880 | 8' 32" | 155 | salt 10'/air | - | 12,79 | 887 |
| 850 | 17' | 60 | oil | 155°C/1 h | 12,38 | 790 |
| 850 | 17' | 60 | oil | 155°C/1 h | 15,21 | 770 |
| 850 | 1h 04" | 155 | salt 10'/air | - | 13,27 | 866 |
| 850 | 1h 04" | 60 | oil | - | 13,59 | 810 |
| 850 | 1h 04" | 60 | oil | 155°C/1 h | 16,36 | 780 |
| 940 | 1' 04" | 155 | salt 10'/air | 180°C/5' | 11,78 | 860 |
| 940 | 1' 04" | 155 | salt 10'/air | 210°C/5' | 15,38 | 815 |
| 925 | 32" | 155 | salt 10'/air | 180°C/5' | 15,17 | 850 |
| 895 | 32" | 155 | salt 10'/air | 180°C/5' | 18,42 | 826 |

obtainable by the classical quench treatment (850°C for 17'/oil) . During subsequent tempering (180-250°C), hardness maintains near "after quenching" values up to 4-5' staying at temperature, whereas a marked decrease occurs for longer times. In this second stage, for times dependent on tempering temperature, hardness decrease below the threshold level of 770 HV obtained by the reference treatment (850°C for 17'/ oil quenching, tempering at 155°C for 1 h). Heat treatments leading to below 770 HV were abandoned.

$K_{IC}$ results (Table 1) indicate multiple possibilities in the choice of time-temperature combinations in heat treatment, that still yield $K_{IC}$ values similar to the ones achievable (15 MPa$\sqrt{m}$) by the reference heat treatment. Fracture morphologies of quenched samples show that brittle intergranular decohesion is prevalent (Figure 2), with areas broken by "complex cleavage". No real difference has been found between medium and low phosphorus samples in fracture toughness (Table 1). In the tempered samples the intergranular areas are less visible but still detectable, with more frequent and pronounced dimples corresponding to the carbides (Figure 3), thus justifying a general decrease in brittleness.

## DISCUSSION

Intergranular brittleness of the type shown in the previous paragraph is known since the early seventies in high carbon steels [5], but several aspects of the phenomenon remain still not fully understood. First of all, an active role of phosphorus in promoting intergranular fracture [6] has to be ruled out, since no real $K_{IC}$ or fracture morphology differences has been found. The explanation offered in [7] appear more convincing: in as quenched high carbon steel, intergranular brittleness is due to the stresses resulting from impingement of growing martensitic plates in adjacent austenitic crystals, sometimes leading to microcracking during quenching. In our case, even if no microcracking was detected in untested as-quenched samples, the explaination seems to hold and may justify the occurrence of weakened grain junctions. The fact that no microcrack was present in the as quenched untested condition may be related to the staying for 10' below the point $M_s$, which results in a gradual martensitic transformation. Moreover, it should be considered that the impingement stress is directly related to the martensite plate length, which in turn is controlled by the prior austenite grain size. Since for SAE 52100 steel during austenitization in different conditions the growth of the grains is limited [4], we can expect that, for all heat treatment combinations proposed here, the austenite grain size be in the narrow range of 7-13 um. A prior grain diameter substantially constant implies the same intergranular brittleness, as confirmed by the results reported in Table 1. It may be concluded that as-quenched fracture toughness of SAE 52100 is quite insensitive to the austenitization conditions, provided that during the subsequent cooling an isothermal permanence below $M_S$ be realized.

The alternative austenitization treatments may then be performed without reducing mechanical properties. As shown in Table 1, the following tempering stage allows to recover a modest degree of toughness, and the results of the alternative heat treatments seems to indicate $K_{IC}$ values in general equal or higher than the classical one (15 MPa$\sqrt{m}$ corresponding to 850°C for 17', oil quenching, tempering at 155°C for 1 h).

## CONCLUSIONS

Hardness tests, fracture toughness tests and fractographic investigations performed on innovative heat treatment on rolling bearing SAE 52100 steel led to the following results:

- fracture toughness is quite insensitive to the austenitization conditions, because the fracture mechanism in the as-quenched steel is always intergranular, but isothermal cooling below $M_S$ must be carried out to avoid quenching microcracking;
- by higher austenitization temperature and shorter times, more carbon is brougth into solid solution, leading to higher as-quenched hardness
- by short high temperature tempering it is possible to reach an optimum hardness-toughness combination, sligthly superior to the classical one.

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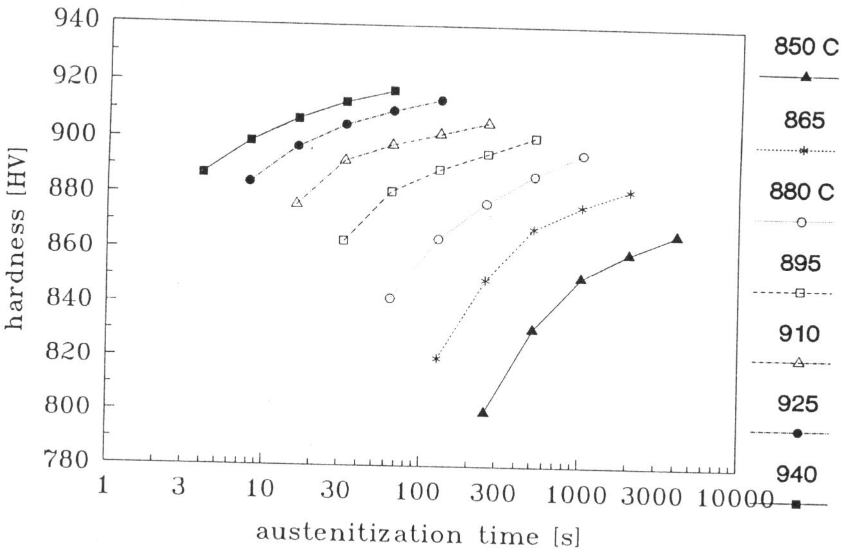


Figure 1. Hardness (HV) versus austenitization time (t) curves for each temperature.

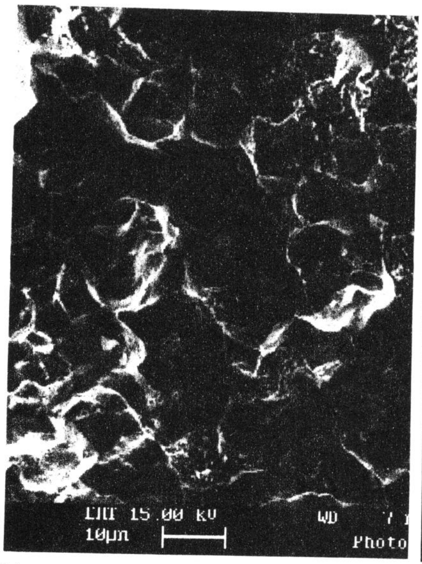


Figure 2. As quenched steel (895°C/1'04", salt 155°C/10').

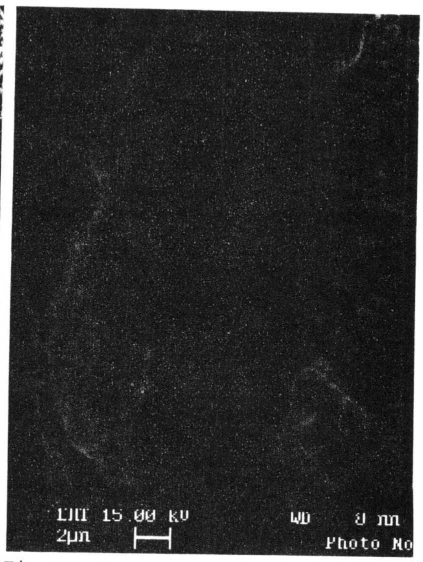


Figure 3. Tempered steel (895°C 1'04", salt 155°C/10', 210°C/5').