The influence of the nitriding temperature on the fatigue limit of 42CrMo4 and En40B steel

J. J. BRAAM, A. W. J. GOMMERS, S. VAN DER ZWAAG
Delft University of Technology, Laboratory of Materials Science, Rotterdamseweg 137, 2628 AL Delft, The Netherlands

The positive effects of surface treatments like nitriding, carburizing, induction hardening and shot-peening on the fatigue properties of construction materials have been known for several decades [1–5]. It has been shown that the improvement in fatigue properties is due to a combination of compressive macrostresses in the outer layer and an improvement of the intrinsic fatigue strength of the material in the layer [6, 7]. The macrostresses are due to a volume expansion of the surface layer caused by the thermomechanical surface treatment. The improvement of the fatigue strength of the surface layer is due to precipitates (nitriding and carburizing) or to a refinement of the microstructure (induction hardening and shot-peening). Usually the influence of macro- and microstresses cannot be separated easily as both effects are the result of the same treatment.

Although the reasons for the improvement in fatigue strength are known, there is little information on the quantitative relations between process parameters of the surface treatment and the fatigue strength. This paper describes some of these studies for gas-nitriding. During gas-nitriding two layers can be formed: the compound layer, consisting of iron nitrides on top of the diffusion layer, consisting of an iron matrix containing precipitates of alloying element nitrides. This diffusion layer is responsible for the improvement in fatigue strength, as the hard compound layer is very brittle and has little mechanical strength. Generally, the fatigue limit increases with increasing diffusion layer thickness, but the relation is non-linear and depends on the local stress concentration.

If the nitriding potential is sufficiently high, the thickness of the diffusion layer depends for a given steel on nitriding temperature and time only. In industrial practice nitriding is invariably performed at the highest temperature possible (580 °C) in order to minimize the process cycle time. However, as nitriding is essentially a precipitation process, nitriding at a high temperature does not automatically lead to the precipitate size and density required for the highest fatigue resistance. Nitriding at lower temperatures could lead to finer precipitates and a better fatigue resistance. Hence, the effect of nitriding temperature on the fatigue limit is examined in this work. The nitriding times at each temperature were adjusted to obtain a diffusion layer thickness of 300 μm in all cases. The nitriding experiments were performed at high nitriding potentials such that a compound layer develops for all nitriding temperatures used. By controlling the composition and the thickness of the diffusion layer the macrostresses are equal, to a first approximation, and the effect of the intrinsic fatigue strength of the nitrided layer can be studied.

In this research, two types of steel have been used: En40B and 42CrMo4. Table I lists the chemical compositions. Before nitriding the steels were quenched and tempered at a temperature of 600 °C for 1 h. The specimens were nitrided in a retort oven (Smit Ofenbau type SGS 30/60). Pure ammonia was introduced into the oven and a dissociation of 30% was reached at the outlet, while the flux was maintained at 10 l min⁻¹. At 450 °C a dissociation of only 20% could be reached. Both steels were nitrided at three temperatures, namely 450, 510 and 570 °C, and the temperature was kept within a range of 2 °C. Table II shows the time-temperature combinations employed to obtain a constant nitriding depth of 300 μm. Before the specimens were nitrided they were cleaned with acetone to remove grease and etched with nitral 5% to prevent non-nitridable spots on the surface.

For the fatigue experiments rotation-bending tests were applied using 4-point bending. The angular

<table>
<thead>
<tr>
<th>Steel</th>
<th>Nitriding temperature (°C)</th>
<th>Nitriding time (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>En40B</td>
<td>450</td>
<td>96</td>
</tr>
<tr>
<td></td>
<td>510</td>
<td>32</td>
</tr>
<tr>
<td></td>
<td>570</td>
<td>8</td>
</tr>
<tr>
<td>42CrMo4</td>
<td>450</td>
<td>43</td>
</tr>
<tr>
<td></td>
<td>510</td>
<td>18</td>
</tr>
<tr>
<td></td>
<td>570</td>
<td>6</td>
</tr>
</tbody>
</table>

Table I: Chemical composition of the steels En40B and 42CrMo4

<table>
<thead>
<tr>
<th>Steel</th>
<th>%C</th>
<th>%Si</th>
<th>%Mn</th>
<th>%P</th>
<th>%S</th>
<th>%Cr</th>
<th>%Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>En40B</td>
<td>0.20–0.28</td>
<td>0.10–0.35</td>
<td>0.45–0.70</td>
<td>0.035</td>
<td>0.035</td>
<td>3.00–3.50</td>
<td>0.45–0.65</td>
</tr>
<tr>
<td>42CrMo4</td>
<td>0.38–0.45</td>
<td>0.40</td>
<td>0.60–0.90</td>
<td>0.035</td>
<td>0.015–0.050</td>
<td>0.90–1.20</td>
<td>0.15–0.30</td>
</tr>
</tbody>
</table>

0261-8028 © 1997 Chapman & Hall
velocity was 5000 r.p.m. (i.e. 80 Hz) and the experiments were done in air at room temperature. For the fatigue tests, unnotched and notched specimens with a diameter of 6 mm were made from each steel. The notches have a stress concentration factor of about 2. The fatigue limit was determined with the so-called staircase method and was defined at $10^7$ cycles. For each nitriding condition twenty specimens were tested. All specimens used to determine a staircase were nitrided as one batch. The average fatigue limit and its variance were determined according to the ASTM procedure [8]. The hardness profiles have also been determined using Vickers microhardness indentations. Each hardness value is the average of at least ten measurements.

The hardness profiles are shown in Figs 1 and 2. For En40B the profiles are steep and the increase in hardness is large. This can be explained as in En40B most of the chromium atoms are not bound as carbides before nitriding. Therefore, the chromium nitride precipitates that develop during nitriding will be very small and coherent with the iron matrix. As the precipitates are small, their formation is relatively rapid. In 42CrMo4 the chromium atoms in the untreated material are bound as carbides, resulting in larger incoherent chromium nitrides upon nitriding, hence a smaller increase in the hardness. In addition, it takes more time to break down the carbides and form nitrides, which results in a shallower hardness profile.

The fatigue limit of the unnotched specimens is shown in Fig. 3 as a function of the nitriding temperature. For En40B the fatigue limit does not depend on the nitriding temperature. This result can be explained by using the local fatigue strength concept [9]. As the fatigue strength in the core is much lower than that in the case (because of chromium nitride precipitates and macrostresses) and because the applied stress increases linearly with the radius in unnotched specimens, the fatigue crack starts beneath the nitrided layer. This is clearly recognisable on the fractured surface. Therefore, the fatigue limit is determined by the local fatigue strength of the base material, so it will not change if different nitriding conditions are applied (provided that the nitriding depth is kept constant).

For the 42CrMo4 steel the transition from the diffusion layer to the base material is gradual and the fatigue limit is more likely to be affected by the interaction between micro- and macrostresses. It is, however, within the accuracy of these measurements, not possible to decide whether there is a significant effect of the nitriding temperature on the fatigue limit.

In the case of notched specimens the applied stress increases strongly near the surface and therefore crack initiation takes place at the surface, i.e. in the nitrided layer. For both materials there is a significant influence of the nitriding temperature on the fatigue limit as can be seen in Fig. 4. It appears
that the fatigue limit decreases with increasing nitriding temperature. The fatigue limit of En40B depends more strongly on the nitriding temperature than that of 42CrMo4. This can be explained as the chromium nitrides in 42CrMo4 develop from carbides whereas in En40B the chromium nitrides develop from substitutionally dissolved chromium. Therefore, in 42CrMo4 the precipitate size will be determined by the size of the chromium carbides rather than the nitriding temperature. In En40B the size of the precipitates is determined by a diffusional process, which is temperature dependent. Higher temperatures lead to a higher diffusivity and therefore larger precipitates can develop. The stronger dependence on the nitriding temperature for En40B is also promoted by the higher chromium content. The effects described can be verified with the hardness profiles. For En40B, the maximum hardness in the case is lower after nitriding at a higher temperature because of the larger precipitate size, whereas for 42CrMo4 the surface hardness is independent of the nitriding temperature. It should be pointed out that the observed increase in fatigue limit with decreasing temperature is contrary to that expected based on thermally induced residual macrostresses. Hence, the increase in fatigue limit recorded is due to an improvement of the intrinsic fatigue properties of the diffusion layer.

In conclusion, nitriding at lower temperatures has no effect on the fatigue limit of unnotched specimens but increases the fatigue limit of notched specimens considerably. The increase seems to depend on the content of substitutionally dissolved chromium in the steel. The results obtained can be explained qualitatively using the local fatigue strength concept.

Acknowledgement
The financial support of the Foundation for Fundamental Research of Matter (FOM) is gratefully acknowledged.

References

Received 24 February
and accepted 21 April 1997