Accelerated carbide spheroidisation of 1.2343 tool steel by induction heating

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Abstract. Tool steels undergo spheroidisation or soft annealing to enhance machinability and cold formability. Conventional soft annealing takes several hours. The final microstructure is composed of globular carbides in a ferritic matrix. We present an alternative process of carbide spheroidisation and steel softening. Accelerated carbide spheroidisation and refinement (ASR) was achieved by induction heating at temperatures close to the $A_1$ temperature. The spheroidised structure was obtained in less than 5 minutes. The carbide particles that formed during the ASR were significantly finer than for the conventional soft annealing. The hardness after ASR was higher than the hardness after soft annealing because of the dispersion strengthening by finer and more densely distributed carbide particles. On the other hand, the fine structure is favourable for hardening. It enables smaller austenite grains and martensite laths to be obtained.

1. Introduction
Manufacturing routes for products from tool steels often include soft annealing, typically before machining [1]. Upon metallurgical processing, alloyed tool steels have pearlitic, bainitic or martensitic microstructures, depending on their chemical composition and the cooling rate after casting or hot forming. As these microstructures are not favourable for machining or cold forming to the final shape, soft annealing is carried out. The steel is heated to approximately the $A_1$ temperature and held there for several hours. Carbon diffusion leads to carbide spheroidisation driven by the need to minimise the ferrite-carbide interface area. The final step in soft annealing is slow cooling in a furnace in order to produce a microstructure composed of a soft ferritic matrix and globular carbides.

1.2343 tool steel is widely used for hot work moulds, dies or punches [2]. Its conventional thermal treatment route comprises soft annealing at approximately 800°C, heating and subsequent quenching from 1050°C, and tempering. Soft annealing is crucial to the material’s machinability and has implications for the quenching schedule. Coarsened spheroidal carbides enriched with alloying elements need to be held at a relatively high austenitizing temperature for a considerable time (typically 1 hour) in order to dissolve. Austenite grain growth is controlled by the pinning effect of vanadium carbides, which remain stable up to 1100°C. The austenite grain size dictates the maximum size of the martensite crystals after quenching. Finer microstructure possesses better material performance [3-5].

The presented article is devoted to the process of accelerated carbide spheroidisation and

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refinement (ASR). The aim is to obtain a microstructure morphology identical to the soft-annealed microstructure with finer and more densely distributed carbides in the ferritic matrix. The ASR process used as a substitute for lengthy soft annealing should take no more than 10 minutes [6]. Two positive effects are sought: a reduction in the processing time and better performance characteristics of the steel. Bainitic and martensitic structures in the initial condition are favourable for obtaining fine globular carbides [7], especially in case of steel alloyed by Mo and V [8].

The ASR process consists in rapid heating to a temperature around the \( A_1 \) and may involve cycling around that temperature. The cycling causes repeated austenitization and divorced pearlitic transformation [9]. It is therefore particularly effective for the induction heat treatment of small products, in which rapid temperature changes can be induced. It also permits switching from batch processing (soft annealing in a furnace) to continuous processing, which is much more suitable for small series. Finer globular carbides lead to finer structures after quenching. Densely dispersed vanadium carbides retard the austenite grain growth more effectively than the coarse conventionally soft-annealed microstructure. On the other hand, smaller Cr and Mo carbides dissolve more rapidly. To obtain the same level of carbide dissolution, the fine microstructure can be austenitized more rapidly or at a lower temperature than a coarse microstructure.

2. Experimental

2.1. Material
The experimental material was 1.2343 tool steel with the chemical composition shown in table 1.

| Element | C  | Si  | Mn  | Cr  | Mo  | V  | Ni  | Cu  | Al  | S  | P  |
|---------|----|-----|-----|-----|-----|----|-----|-----|-----|----|----|
| wt. %   | 0.40 | 0.96 | 0.35 | 4.98 | 1.19 | 0.42 | 0.22 | 0.11 | 0.03 | 0.003 | 0.024 |

Table 1. Chemical composition of experimental steel.

The steel was in the form of hot-forged bars 19 mm in diameter. The bars contained upper and lower bainite and a small amount of martensite (figure 1). The hardness of the material in the initial state was 538 HV10.

Figure 1. Microstructure of the material in its initial state: a) optical micrograph, b) scanning electron micrograph.

2.2. Heat treatment
The samples were treated in a quenching dilatometer Linseis L78 RITA, in which cylindrical samples with a diameter from 3 to 5 mm and a typical length of 10 mm can be tested. The present samples were heated by induction and cooled by gas flow. The available controlled heating and cooling rates are up to 200°C·s\(^{-1}\). The temperature range is from -200°C to 1400°C. The chamber of the dilatometer
can be evacuated or filled with an inert atmosphere. The cylindrical samples used for heat treatment in the dilatometer were 4 mm in diameter and 10 mm in length. These dimensions permitted hardness measurement, microstructure evaluation, as well as micromechanical testing of the samples [10].

The heat treatment schedules comprised the following steps: heating at a rate of 35°C·s⁻¹, holding at the chosen temperature and subsequent gradual cooling at various cooling rates. Cooling from the holding temperature to 670°C took place at a rate of 1.97°C·s⁻¹. Between 670°C and 500°C, the cooling rate was 1.11°C·s⁻¹. Between 500°C and 400°C, it was 0.71°C·s⁻¹. Final cooling to ambient temperature took place at a rate of 10°C·s⁻¹. The heating and cooling rates were chosen to simulate induction heating and air cooling of a rod 15-20 mm in diameter. Specimens of this size are intended to be used in the next stage of the experimental programme. The samples in the first series were rapidly cooled from 700°C to ambient temperature. The purpose of this rapid cooling from 700°C was to preserve the microstructure characteristics present at that temperature, namely the austenite which had formed during the treatment and remained stable at temperatures below the A₁. The holding times at the annealing temperatures were chosen at 15, 45 and 150 seconds, based on previous experiments with 100CrMnSi6-4 chromium bearing steel.

The schedule designation codes contain information about holding times and temperatures. Where rapid cooling from 700°C was used, the schedule code ends with the suffix “q” (figure 2).

![Figure 2. Time-temperature diagrams of heat treatment schedules. a) Schedules designated as “850°C/15s” and “850°C/15s-q” with rapid cooling from 700°C; b) Schedule “850C/15s 675C/180s”.](image)

2.3. Examination of samples
A dilatometric curve was recorded for each specimen in order to identify phase transformations during the ASR process. After the treatment, a longitudinal metallographic section through each sample was prepared. These sections were ground and polished. HV10 hardness was measured along the sample axis. The microstructure was revealed by etching with nital and observed under a Nikon Eclipse 200D optical microscope and JEOL JSM 7400F scanning electron microscope.

3. Results and discussion

3.1. Hardness and dilatometric measurement
The first stage of the experimental programme was focused on carbide coarsening in ferrite and austenite formation. The samples were heated to and held at temperatures between 730°C and 870°C for 15 seconds. Afterwards, they were gradually cooled to 700°C and then cooled rapidly at 200°C/s. The rapid cooling was employed to find the amount of austenite that had formed during the 15-second holding time. This austenite would transform to martensite during cooling. The hardness of the samples after these schedules is shown in table 2.

The final hardness decreases with increasing temperature of the 15-second holding period. This is due to the precipitation of carbides from the martensite and the spheroidisation and coarsening of the bainitic carbides. The significant increase in hardness after schedules with holding temperatures of
850°C and 870°C is caused by austenitization and martensitic transformation during rapid cooling.

**Table 2.** Hardness levels after schedules with 15-second and 50-second holding times and with quenching from 700°C. The letters IS stand for “initial state”.

| Schedule        | Hardness HV10 | Schedule        | Hardness HV10 |
|-----------------|---------------|-----------------|---------------|
| IS              | 538           | ...             | ...           |
| 730C/15s-q      | 497           | ...             | ...           |
| 750C/15s-q      | 438           | ...             | ...           |
| 770C/15s-q      | 391           | ...             | ...           |
| 790C/15s-q      | 362           | 790C/150s-q     | 303           |
| 810C/15s-q      | 332           | 810C/150s-q     | 286           |
| 830C/15s-q      | 319           | 830C/150s-q     | 292           |
| 850C/15s-q      | 343           | 850C/150s-q     | 353           |
| 870C/15s-q      | 473           | ...             | ...           |

The holding time was then extended to 150 seconds in those schedules which initially led to the lowest hardness values (table 2). The hardness behaviour was the same as with the 15-second holding time. The minimum hardness value of 286 HV10 was obtained with the 810C/150s-q schedule. The 830C/150s-q schedule led to a virtually equal hardness of 292 HV10.

As expected, higher temperatures and longer holding times at a particular temperature are more effective in softening the material. The temperature limit for softening is the temperature of austenite formation. The presence of austenite can be favourable for rapid spheroidisation of carbides, as it facilitates the dissolution of small carbide particles in bainite or those that precipitated from martensite during heating. However, austenite has to undergo divorced pearlitic transformation during cooling. Otherwise, lamellar pearlite, bainite or martensite form during cooling and softening of the material is reversed.

The purpose of the schedules without quenching, 850°C/15s, 850°C/150s and 870°C/15s, was to determine whether austenite decomposes during gradual cooling of the sample. 850°C/15s and 830C/15s-q were the only schedules which led to lower hardness values without any austenite being formed. Metallographic analyses and dilatometric plots did not reveal any lamellar pearlite, bainite or martensite in the microstructure. Thus, it can be assumed that divorced pearlitic transformation took place during gradual cooling. Other samples listed in table 3 showed higher hardness and a small amount of martensite in the microstructure.

**Table 3.** Hardness of specimens upon schedules with gradual cooling to ambient temperature.

| Schedule        | Hardness HV10 | Schedule        | Hardness HV10 |
|-----------------|---------------|-----------------|---------------|
| 850C/15s        | 313           | 850C/150s       | 333           |
| 870C/15s        | 415           | ...             | ...           |

The final stage of the experimental programme was devoted to estimating the austenite stability. The schedules used at this stage involved austenitizing at temperatures of 850°C and 870°C and holding for 1800 s at 675°C (table 4). This holding time was chosen in order to ensure that the austenite fully decomposed. Dilatometric analyses showed the austenite decomposition times. The austenite that formed during the 15-second holding time at 850°C took 150 seconds at 685°C to decompose (counted from the holding start). The austenite that formed at 870°C was far more stable and decomposed for 1100 seconds. As this long decomposition time was not acceptable for the ASR process, the maximum temperature chosen for the final experiments was 850°C.

Table 4 shows hardness values after schedules involving partial austenitization and interrupted cooling. The austenitizing times at 850°C were 15, 45 and 150 seconds. The cooling was interrupted by 180-second holding at 675°C. The partial austenitization followed by divorced pearlitic transformation did not contribute to further carbide coarsening when compared to the 810C/150s-q schedule.
Table 4. Schedules with the hold at 675°C for austenite decomposition.

| Schedule          | Hardness HV10 |
|-------------------|---------------|
| 850°C/15s 675°C/1800s | 299           |
| 870°C/15s 675°C/1800s | 294           |
| 850°C/15s 675°C/180s  | 308           |
| 850°C/45s 675°C/180s  | 298           |
| 850°C/150s 675°C/180s | 312           |

3.2. Metallographic analysis
All samples contained carbides of similar morphology in the form of rounded or edged particles. Their size varied from approximately 20 nm to 300 nm. Carbides coarsened with increasing temperature of the induction treatment. The coarsening was visible in both the smallest and the largest carbides. The typical carbide size in the 750°C/15s-q sample was between 20 nm and 150 nm (figure 3(a)). In the 810°C/15s-q sample, it was between 30 nm and 200 nm (figure 3(b)). The lowest hardness was achieved in the 810°C/150s-q sample: 286 HV10. The microstructure retained the differences between the bainitic and martensitic regions visible at lower magnifications (figure 4(a)). However, carbides were homogeneously distributed in the ferritic matrix. They were coarsened, with sizes up to 300 nm (figure 4(b)).

Figure 3. Microstructures upon schedules: a) 750°C/15s-q, b) 810°C/15s-q.

Figure 4. Microstructure upon schedule 810°C/150s-q: a) optical micrograph, b) scanning electron micrograph.
The lowest hardness achieved in the experimental programme was around 290 HV10. The hardness after conventional soft annealing is usually under 230 HV. From the morphological point of view, the ASR process produced an equivalent but much finer microstructure than soft annealing. In particular, the finest carbides below 100 nm in diameter are those which are probably responsible for the dispersion strengthening. There are two ways to ensure that coalescence of small carbides and the larger ones occurs. One mechanism is Ostwald ripening induced by annealing below the $A_t$ temperature. This is the conventional way used in soft annealing. The second approach involves austenitizing the ferritic matrix, dissolving the finest carbides and inducing divorced pearlitic transformation during cooling. If the ASR process is to be applied, the treatment should be as brief as possible. Rapid austenitization is therefore a suitable option.

Austenitization took place in schedules with heating at temperatures above 830°C. The divorced pearlitic transformation occurred during holding at 675°C. However, austenitization at 870°C for 15 seconds was not sufficient to dissolve even the finest carbides. The resulting size and distribution of carbides were the same as in schedules without austenitization. Future experimental work devoted to further softening of the steel should focus on fine carbide dissolution. This will entail finding the optimal temperature for austenite decomposition via divorced pearlitic transformation.

If a hardness of 290 HV is sufficiently low for a particular production process, then the schedules described in this paper could be used. Further work must be devoted to hardening of the fine-structured material obtained from the ASR process. The fine martensitic structure promises better performance of the final product.

4. Conclusion
Induction heat treatment of 1.2343 steel was performed. The initial state with a bainitic-martensitic structure underwent spheroidisation annealing for less than 5 minutes. A microstructure of globular carbides in a ferritic matrix was obtained. The most effective route for carbide spheroidisation and coarsening was annealing just below the $A_t$ temperature. After holding for 150 seconds at 810°C, the hardness dropped from the initial 538 HV10 to 286 HV10.

Rapid austenitization at temperatures of 850°C or higher, with subsequent divorced pearlitic transformation, led to similar results. Optimization of the austenitization and austenite decomposition cycle will be one of the aims for further research.

The process of accelerated carbide spheroidisation produces much finer globular carbides than conventional soft annealing. The influence of the microstructure fineness on the material’s performance after hardening will be the subject of the next stage of the experimental programme.

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