Warm Forged Medium Carbon V Steel

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Abstract: Nowadays there is a continuous demand, particularly from the automotive industry, for cheaper, lighter and more reliable components. It is not surprising then that steel research has been focused during the last decades in new qualities and processes. This paper is dealing with the use of vanadium microalloyed steels on one of those new processes, warm forging. For its low precipitation temperature and its recognised ability to strengthen steel microstructures via austenite grain growth control, precipitation hardening and interference of the static recrystallization process, vanadium in microalloyed steels seem to be an appropriate candidate for warm forging.

Key words: Warm forging; vanadium microalloyed steels.

During the last 25 years, the steel industry has experienced some of the most important changes at two different fronts, steel design and processes, the driving force being the continuous demand for cheaper, lighter and more reliable components. In this sense steel manufacturing industries are pushing towards the production of components that exhibit dimensions that are more precise, need less machining and require less part processing. It is worth emphasising that the distribution of the costs in producing conventional forging components shows that more than 50% is assigned to machining [1]. In consequence, a great interest has emerged in the near-net-shape technologies, as cold and warm forging processes.

Warm forging (in the temperature range from 600 to 900ºC) has a number of advantages over traditional forging procedures [2]. Among them, better utilisation of material, improved surface finish, dimensional accuracy compared with hot forging, reduced press loads compared with cold forging, and control of the microstructure such that the desired properties can be obtained without further heat treatments. The introduction of microalloyed steel grades in warm forging has been considered as an option to obtain better strength-toughness combinations in comparison to plain C-Mn steels [2,3].

In the case of warm forging of V steels, it is necessary to keep in mind that depending on the reheating temperatures prior to deformation, different quantities of V(C,N) may remain undissolved, interacting with the microstructure in different scenarios, i.e. controlling austenite grain size growth and similarly, delaying austenite static recrystallisation kinetics after deformation [4]. The V that is in solid solution is available for further precipitation during cooling after deformation.

The aim of this work is to study the potential that warm forging in combination with V microalloying may have on the final properties of steels. For this purpose a vanadium microalloyed steel grade has been used.

1 Experiment Method

A commercial vanadium microalloyed steel containing, all in wt.%) 0.24C-1.56Mn-0.28Si-0.10Cr-0.24Cu-0.18V-0.01N has been selected for the purpose of this work. Warm forging simulations were performed in a servo hydraulic press by testing plane strain compression specimens of 25x60x10 mm³, machined from industrially hot rolled steel bars. Specimens were heated and soaked at the deformation temperature for 10 min before being deformed at a strain of $\varepsilon = 0.3$ and strain rate of 10 s⁻¹. Following deformation, specimens were subjected to two different cooling rates, i.e. air cooling ($\approx 1^\circ$C/s) and accelerated cooling ($\approx 4^\circ$C/s). Three different testing temperatures were considered 800, 835 and 870ºC, all of them in the austenite range.

Prior austenite grain size (PAGS) before deformation was determined metallographically from quenched specimens after soaking at the testing temperature. Its size was measured by the mean equivalent diameter method.

Microstructural characterisation was carried out using optical and scanning electron microscopy on metallographic samples cut longitudinally from the plane strain compression specimens. Round tensile samples (Ø = 4 mm and L = 10 mm) were machined from plane strain compression specimens.

Carbon replicas were prepared by conventional methods and observed in a transmission electron microscope fitted with an energy dispersive X-ray system. The size and distribution of V precipitates were quantified measuring in each case a minimum of 500 and a maximum of 1200 particles.

2 Results and Discussion

2.1 Microstructure before deformation

PAGS measurements reveal that it remains almost constant with an average size of about 5.5 µm, between
835-870°C, a clear indication that undissolved V precipitates at those temperatures are exerting a pining effect in the austenite grain boundaries, impeding its advance. This is supported by the fact that measurements performed at higher temperatures, 900 to 1100°C showed an increment of the PAGS from 8 to 83 µm respectively, as all the V comes into solid solution. Similarly, the size distributions of precipitates measured in the samples quenched after reheating at 835 and 870°C, as it will shown later, confirm that they remain very fine and that, in consequence, are able to exert a control in the austenite grain size.

The other microstructural aspect to take into account before deformation is applied is the situation of V(C,N) precipitates. In the as received (AR) condition, the spherical particles observed were identified as V rich precipitates. In Fig.1 it can be seen a distribution of very fine precipitates with a mean diameter of 8 nm, 71% of the precipitates being smaller than 9 nm. The number of particles per area unit (ρ) is also indicated in the figure. Fig.1 also shows the size distribution of V precipitates after reheating for 10 min. at 870 and 835°C followed by water quench, identified as AR-TºC. If these results are compared with those of as received condition, it is possible to see a displacement towards coarser particle sizes as heating temperature increases. This denotes that there has been a dissolution process involving, above all, the smallest precipitates, which is also confirmed by data in ref [5]. Although it is not probable due to the moderate reheating temperatures, it could be argued if coarsening of the size distribution is assisted or not by Ostwald ripening. Using Wagner formalism [6], the time required for the coarsening of V precipitates by Ostwald ripening was calculated. The results indicate that a particle of 6 nm needs 1.8 hours and 7 hours to achieve 10 nm at 870°C and 835°C, respectively. This means that coarsening by Ostwald ripening mechanism is very improbable to occur with the applied treatments, thus supporting the idea that the larger size of precipitates after reheating compared to the as received condition may exclusively arise from the dissolution of the smallest particles, as mentioned above. According with the latter discussion, it is reasonable to assume that at 800°C, where small dissolution has taken place, the size distribution and density of particles must be very close to that of the as received conditions, i.e. higher density of smaller particles as compared to 870 and 835°C reheating temperatures.

2.2 Microstructure after deformation

The microstructure at room temperature consists of a mixture of ferrite, pearlite and martensite/austenite constituent (MA), as shown in Fig.2. The pearlite present has not its normal lamellar structure and can be considered as “degenerated pearlite”.

Fig.1. Characterization of V(C,N) precipitates. AR = As Received, AR-TºC = After Reheating at T and water quench, AD-TºC = After Deformation at T and air cooling.

Fig.2. SEM micrographs of the generated microstructure after deformation at 870°C followed by (a) air cooling and (b) accelerated cooling.

The metallographic measurements in terms of ferrite and MA mean grain size and volume fraction are shown in Fig.3 as a function of the deformation temperature. As observed in the figure, some room temperature microstructural features are very dependent on the deformation temperature. This temperature can interact with the transformation during cooling by modifying two factors: the austenite grain conditioning before transformation and the V dissolution/precipitation situation. Concerning the austenite microstructure before transformation, in a previous work [1] it was quantified that the presence of undissolved V(C,N) precipitates retards the static
recrystallisation process occurring after deformation in the warm temperature regime. This delay becomes longer for lower deformation temperatures. Taking into account this, it can be concluded that in the samples tested at 870°C, before transformation begins, there will be sufficient time to achieve a completely recrystallisation of austenite. In contrast, at 835 and 800°C the recrystallisation will be incomplete. This implies that the austenite grain boundary area per unit volume, $S_v$, before transformation will increase in these last two conditions. It is well known that higher $S_v$ values favour the microstructure refinement during transformation. This refinement effect is enhanced as the cooling rate is increased. In this context, the initial austenite grain size prior to deformation is very small, and, as a consequence, the resulting $S_v$ value in the partially recrystallised microstructures deformed at 800 and 835°C will be high enough to bring about a very fine ferrite grain size, as shown in Fig.3, with a minor effect of the cooling strategy. This can explain the behaviour of the ferrite grain size remaining nearly constant at 800 and 835°C. In contrast, as at 870°C the austenite microstructure is completely recrystallised before transformation, $S_v$ will be smaller than in the previous cases, thus leading to some refinement in the ferrite mean grain size when accelerated cooling is applied.

The dependency of the ferrite fraction with deformation temperature can be considered together with the tendency of increasing MA fraction (Fig.3). Different factors can be interacting simultaneously. Firstly, the accumulated strain in the austenite (at low deformation temperatures) will accelerate the ferrite transformation, i.e. ferrite will start forming at higher temperatures and shorter times, thus favouring higher ferrite fraction contents \([7]\). Secondly, V in solution moves the end of pearlite transformation to longer times \([8]\). These two factors can explain qualitatively the behaviour observed where a decrease in the ferrite and an increase in MA volume fractions occur as deformation temperature increases.

Finally, the precipitate size distributions after deformation followed by air cooling, AD-T°C, are shown in Fig.1 at deformation temperatures of 800 and 870°C. These results indicate that deformation at 870°C induces a refinement in the particle size distribution while lower deformation temperatures have the opposite effect. The possibility of strain induced precipitation in austenite can be ruled out because of the absence of a plateau denoting this effect in the recrystallised fraction versus time plots reported in ref. 4. Therefore, all the V in solution will be available for further precipitation during cooling, either as new precipitates or on already existing ones.

A higher reheating temperature will lead to an increase in the precipitation rate of new particles through a higher supersaturation during cooling but, on the other hand, lower dissolution temperatures will favour accelerated precipitation of the solutes on pre-existing particles, which act as preferred sites \([8]\). And this is essentially the situation detected. When comparing data of Fig. 1, it is clear that a refinement in the particle size distribution is attained after deformation applied at 870°C. The mean size decreases slightly from 21 to 18 nm and there is an important increase in the fraction of particles smaller than 9 nm. This implies that a fine fresh precipitation is taking place during subsequent cooling after deformation, resulting in an important increase in the density of precipitates detected i.e. from 11 to 91 $\mu$m$^2$. This situation differs from that observed in the samples deformed at 800°C. Assuming that before deformation the precipitation state at 800°C should be very similar to that of the as received condition, there is an increase in the average precipitate size from 8 to 14 nm, in addition to very similar values of $\rho$ and an important increase in the amount of particles bigger than 17 nm. This suggests that the small amount of solute vanadium available is precipitating mainly on pre-existing particles.

![Fig.3. Quantitative metallography of ferrite and MA constituent as a function of warm forging conditions.](image)

### 2.3 Mechanical properties

Concerning the room temperature mechanical behaviour, summarized in Fig.4, a small increase in the measured UTS is detected as the deformation temperature and cooling rate increases, the later leads to an increase of almost 170 MPa at all temperatures. On the other hand, YS tends to increase as temperature increases for air cooling samples while the opposite behaviour is given in the case of accelerated cooling.

Ductility, measured as the reduction in area, increases as forging temperature decreases in the V microalloyed steels, showing higher values for the air cooled than for accelerated cooled.

The multiphase microstructure developed made difficult the quantification of the contribution of each hardening mechanism to the overall strength. Nevertheless, there are some general rules that can be...
taken into account. For example, the influence of the ferrite, fraction and size, on the YS is evident when comparing data in Fig.3 and Fig.4, i.e. decreasing as smaller fractions and larger grains of ferrite are present at the microstructure.

![Fig. 4. Mechanical properties as a function of warm forging conditions.](image)

The contribution of other hardening mechanisms, that is, hard constituents as MA (Fig.4), precipitation strengthening and dislocation density (mainly in the accelerated cooled samples) should be taken into account. Thus, MA fraction seems to be controlling the UTS and ductility behaviour. Higher fractions of smaller MA constituents increase the UTS and decrease the reduction in area. Precipitation hardening evidence has been presented in ref. [10], where nano-hardness measurements of ferrite were made after different deformation conditions. The results thus obtained showed that ferrite present at the microstructure becomes softer as the deformation temperature and subsequent applied cooling rate decrease. The results are rationalized in terms of the effectiveness of the precipitation strengthening contribution, i.e. precipitate size and distribution.

3 Conclusions

The microstructural and mechanical behaviour of a V microalloyed steel showed the importance of the different amounts of V in solution, which depends on the applied reheating temperature before warm deformation.

Given the low temperatures implied in warm forging procedures, i.e. low levels of dissolution and limited coarsening by Ostwald ripening, the size and distribution of the vanadium precipitates in the as received condition are of great importance. These undissolved particles can play different roles: control of the austenite grain growth during reheating and delay of austenite static recrystallisation after deformation.

Microalloying with V leads to fine multiphase microstructures formed by ferrite, pearlite and MA phases. Depending on the reheating temperature, the vanadium in solution will precipitate during cooling as fine particles or precipitate over existing undissolved particles. The amount of each constituent is affected by the reheating temperature.

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References:
