1. Introduction

The application of thermomechanical processes to eutectoid steels is a relatively recent development compared to the amount of research and industrial applications carried out on other steel grades. One of the reasons for this difference comes from the fact that, for a wide range of applications of eutectoid steels, the main microstructural objective is to obtain pearlite with fine interlamellar spacing. This refinement is considered as the best condition for ulterior wire drawing operations and also because the interlamellar spacing is the main factor controlling the strength of the pearlite.\(^1\)\(^,\)\(^2\) In addition to that, eutectoid steels are used in the transport industry for applications that require good combinations of strength, wear resistance and toughness. For these applications, thermomechanical processes have been explored in order to obtain improved strength–toughness combinations, compared to those achieved following conventional routes.\(^3\)

In addition to the refinement of interlamellar spacing via control of the cooling rate or undercooling (in isothermal treatments) during transformation, solid solution and precipitation hardening are the other main treatments that are used to increase the strength of pearlitic steels. In the case of precipitation hardening, microaddition with vanadium is one of the most common approaches.\(^4\)\(^,\)\(^5\)

Regarding toughness, several authors have proposed that the brittle fracture process in pearlitic steels is controlled by a structural unit made up of a number of colonies exhibiting close crystallographic orientations and that this unit is in turn controlled by the prior austenite grain size.\(^6\)\(^,\)\(^7\) Pickering et al. observed that the pearlite morphology gradually changes from a multic colony nodular structure for a coarse prior austenite grain size to a structure composed of individually formed colonies for a sufficiently small prior austenite grain size.\(^8\) The same authors observed that in the nodular structure the cleavage cracks propagate throughout entire nodules, producing flat or stepped facets and stopping or changing direction mainly at the nodule boundaries. On the other hand, the cleavage facets in the colony structure seem to be formed by only one or very few colonies; in this case the obstacles for crack advance are the boundaries between colonies.\(^9\)

Independent of the morphology, it is obvious that the microstructural unit controlling the fracture toughness of pearlite must be directly related to the set of colonies enclosed by high angle boundaries. Consequently, to establish this relationship a limiting angle between high and low angle boundaries must be previously fixed or defined. Park and Bernstein studied the orientation of the pearlitic ferrite planes of adjacent colonies that cleavage cracks propagate across.\(^10\) These authors observed that for the case of a large prior austenite grain size the misorientations between the \{100\} ferrite planes were less than 10° in 90% of the colonies studied. This suggests that a limiting angle larger than 10° must be assumed for a change in the crack direc-

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**Influence of Thermomechanical Processing on the Austenite–Pearlite Transformation in High Carbon Vanadium Microalloyed Steels**

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tion. This means that when defining the “ferrite unit” the misorientation criterion used is an important factor. A 15° misorientation criterion was used in Ref. 11) for an eutectoid Nb microalloyed steel, whereas in Ref. 12), where several V microalloyed steels were studied, it was observed that the application of the above criterion led to an “orientation unit” mean size larger than the “facet size” measured on fracture surfaces. In the latter a 12° misorientation criterion was used.

In a recent paper Jorge-Badiola et al. evaluated the beneficial effect of combining strain induced vanadium nitride precipitation in austenite with proper deformation schedules to accumulate strain in the austenite in eutectoid steels.13) As occurs in low carbon steels, the accumulation of strain produces a significant refinement in the austenite, this leads to high values in the specific grain boundary area ($S_g$) which may produce, after transformation, a refinement in the “ferrite unit” in pearlite. In the aforementioned paper the work was mainly focused on analyzing the relevance of VN precipitation on austenite conditioning and the subsequent influence on the size of “ferrite units” obtained after transformation compared with a plain C–Mn steel.

The precipitation of VN in austenite and the subsequent grain size refinement will affect the CCT curve and consequently, the resulting interlamellar spacing as well as the V(C, N) precipitation hardening can be modified for a given cooling schedule. This paper attempts to evaluate the interaction between these factors considering the influence of different vanadium and nitrogen contents in eutectoid steels.

2. Experimental Procedure

The composition of the vanadium microalloyed eutectoid steels used in this study are indicated in Table 1. Steels 5V and 10V were provided in the form of 16 mm diameter bars forged from laboratory casts. Steel 11VN was provided as square blooms forged from laboratory cast obtained by the addition of vanadium after remelting a commercial eutectoid plain C–Mn steel (Steel C–Mn1 in Table 1). Torsion specimens, having a gauge length of 12 mm and a diameter of 6 mm, were machined from the as-received materials. Specimens were deformed by multipass torsion tests at decreasing temperature that followed different deformation schedules in order to produce different austenite microstructures prior to transformation to pearlite. Before deformation the specimens of steels 5V and 10V were re-heated at 1100°C for 15 min, while for the case of steel 11VN a temperature of 1200°C was selected to assure a complete dissolution of vanadium precipitates. The measured torque $\Gamma$ and the twist $\theta$ were converted to von Mises equivalent stress ($\sigma$) and strain ($\varepsilon$) using these equations:

$$\sigma = \frac{3}{2\sqrt{3}} \frac{3.3 \Gamma}{2\pi r^3}, \quad \varepsilon = \frac{r\theta}{L\sqrt{3}} \quad \text{...........(1)}$$

where $r$ and $l$ represent the diameter and the gauge length of the torsion specimen, respectively.

The deformation parameters used for austenite conditioning are given in Table 2. The strain-rate used in all the tests was $\dot{\varepsilon}=1.5$ s$^{-1}$, and the interpass time was varied between 10 and 20 s, depending on the deformation temperature range. During deformation the cooling rate was 1.5°C/s. All the materials were subjected to two types of deformation schedules: in the first one (schedule A) deformation was only applied at the high temperature range (1075–970°C), this led to a recrystallized austenite microstructure; in the second (schedule B), deformation passes were extended to the low temperature range (below 900°C) in order to accumulate strain in the austenite (pancake microstructure). After the deformation, some specimens were water quenched to reveal the austenite state present before transformation. The rest of the specimens were cooled down to room temperature at different cooling rates in the range from 3 to 8°C/s (measured in the temperature range between 650 and 750°C). For the cooling after the deformation, a cooling device using air-compressed was developed. The system was installed at the exit of the deformation unit (similarly to the quenching unit) and was controlled by an electro-valve that allowed the air output pressure to be regulated. In these tests, a constant air-pressure was used during each cooling test, thus the different cooling rates were obtained by applying different pressures.

After cooling to room temperature, the specimens were prepared for hardness measurements as well as metallographic observation and electron backscattered diffraction (EBSD) analysis. The samples were obtained by cutting a flat surface on the specimen parallel to the torsion axis at a distance close to the subsurface ($r^*=0.9r$, $r$ being the outer radius of the specimen). The microscopy samples were then polished following conventional techniques and prepared for optical analysis by etching with a saturated aqueous picric acid so as to reveal the prior austenite grain boundaries and with 2% nital for pearlite. The austenite microstructure was characterized by the grain boundary area per unit volume ($S_g$) measured using the mean linear intercept method. When the grains are equiaxed,14)

$$S_g = 2N_L / l \quad \text{............(2)}$$

where $N_L$ and $l$ represent the mean number of intercepts.
and the mean linear intercept value, respectively. For deformed microstructures, $S_V$ must be a weighted average of the $N_{ij}$ values in each of the three principal directions. For the deformation by torsion, $S_V$ can be calculated as a function of the strain, by using the following equation:

$$S_V = \frac{S_{i0}}{2} \left[ 0.429 (\sqrt{3}e^2 + 1)^{-1} + 0.571 + (\sqrt{3}e^2 + 1) \right] \ldots (3)$$

where $S_{i0}$ and $S_i$ represent the contributions of the austenite grain boundaries to the specific boundary area prior to and after deformation, respectively.

The pearlite interlaminar spacing ($\lambda$) was measured from samples that were etched with picral and examined in the scanning electron microscope (SEM). $\lambda$ measurements were carried out using the linear intercept method with circles proposed by Underwood ($\lambda = 2L/n$, where $L$ is the total length line and $n$ the number of cementite lamellae intersects). For EBSD observations of pearlitic structures, specimens were polished down to 1 $\mu$m and the final polishing was with colloidal silica. The EBSD scans were carried out in a Philips XL30CP SEM equipped with PEGASUS 4000 (TSL-EDAX) including OIM 4000 EBSD. Since the Kikuchi patterns of cementite in pearlite are not resolvable, only diffraction patterns arising from ferrite were considered. The scan step size was 1 $\mu$m length and the total scanned area was about 600$\times$600 $\mu$m$^2$ per sample. The ferritic “orientation unit” was defined using a tolerance criterion of a 12° change in orientation. The mean size of these orientation units, $D_p$ (measured as equivalent diameter) as well as the size distribution in area were quantified. The procedure followed was similar to that described in Ref. 13.

3. Results

Figure 1 shows the stress–strain curves and the corresponding austenite microstructures obtained after the application of the different schedules described in Table 2 with steels 5V and 11VN. After the application of schedule A, a completely recrystallized austenite was achieved with the three V microalloyed steels, as shown in the example of Fig. 1(a) for steel 5V. In contrast, when schedule B was applied elongated austenite grains, denoting the accumulation of deformation, were observed (Figs. 1(b) and 1(c)).

From the microstructural analysis of the quenched samples, the recrystallized austenite grain sizes prior to transformation obtained with schedule A and the corresponding $S_V$ values (Eq. (2)), together with those determined for the unrecrystallized conditions obtained with schedule B (Eq. (3)) are listed in Table 3. The initial austenite grain size, $D_{i0}$, is also indicated in the table. As vanadium content increases, $D_{i0}$ decreases, this effect being more noticeable in...
steel 11VN with the highest N content. This tendency is also observed after the application of deformation schedules A and B. In both conditions the refinement is more significant in steel 11VN.

The Mean Flow Stress (MFS) values calculated from the stress–strain curves of Fig. 1 for schedule B are plotted in Fig. 2 against the inverse absolute pass temperature for the three steels. The change in the slope separates the region where the material recrystallizes completely between passes from the region where recrystallization is total or partially absent. For this particular deformation schedule there are not main differences between the three steels, with a non-recrystallization temperature, $T_{nr}$, close to 905°C.

Figure 3 shows several optical micrographs of the general pearlite microstructure obtained in steel 5V after the application of both deformation schedules and different cooling conditions. For a given cooling rate the refinement of the pearlite microstructure when transformation takes place from unrecrystallized austenite (schedule B) is clearly obvious and the same is observed for the other steels. On the other hand, some refinement also occurred when the cooling rate was increased, above all when the cooling rate was increased from 3 to 6°C/s.

The average measured hardness values are plotted as a function of the cooling rate on a semi logarithmic scale in Fig. 4. For all the steels, a linear relationship is obtained. It can be seen that the hardness increased when the cooling rate increased. Moreover, for a given cooling rate the unrecrystallized austenite shows a lower hardness compared to the recrystallized one in all the materials. On the other hand, steels 10V and 11VN with a higher content of vanadium show higher hardness values than steel 5V at similar cooling conditions.

The microstructural analysis confirmed that martensite and bainite appear in steel 11VN when transformation takes place from recrystallized austenite at cooling rates higher than 6°C/s, which leads to a significant increase in hardness. In contrast, after schedule B, lower hardness values are obtained as shown in Fig. 5. In this figure, the region where fully pearlitic microstructures were obtained with the three steels is shaded. It can be seen that at 6°C/s, in unrecrystallized condition, steel 11VN got a fully pearlitic microstructure, while at 8°C/s some MB (martensite–bainite) features can be identified in the microstructure which increases the hardness.

The values of the interlamellar spacing measured for the different steels and austenite conditions have been represented as a function of the cooling rate in Fig. 6. Only

### Table 3

| Steel | As-reheated Temp. (°C) | $D_a$ (μm) | Schedule A | Schedule B |
|-------|------------------------|------------|------------|------------|
| 5V    | 1100                   | 150        | 81         | 25         |
| 10V   | 1100                   | 124        | 60         | 33         |
| 11VN  | 1200                   | 105        | 37         | 54         |

Fig. 2. Plot of mean flow stress against inverse absolute temperature for different steels and deformation sequences.

Fig. 3. Optical micrographs showing the general microstructure of pearlite in steel 5V after the application of different thermomechanical sequences and cooling ranges: left) schedule A; right) schedule B. Cooling rate is indicated in each figure.

Fig. 4. Mean hardness values of fully pearlitic microstructures as a function of the applied cooling rate (CR).
measurements corresponding to fully pearlitic microstructures have been used in the figures. For a given steel and austenite condition, the interlamellar spacing decreases as the cooling rate increases. Similarly, for a given cooling rate larger values in the interlamellar spacing are obtained for transformation from unrecrystallized austenite, compared to the recrystallized one in all the steels with the difference being very significant for steel 5V.

Figure 7 shows examples of the grain maps obtained by OIM analysis from the microstructures that were produced after the application of A and B deformation schedules to steel 11VN. Similar maps were generated for the different steels and deformation-cooling conditions. From these maps, the “orientation unit” sizes were measured by quantitative metallography methods taking the equivalent diameter as the parameter for measurement. The units separated by boundaries <12° were taken to be the “orientation units” that can become single cleavage facets after fracture. The size distributions were also quantified in all cases and these are represented in Fig. 8. In general, a refinement of the ferrite units is observed when transformation takes place from unrecrystallized austenite (schedule B), compared to the recrystallized one (schedule A), mainly at the lowest cooling rate of 3°C/s. On the other hand, when the cooling rate was increased the ferrite units were also refined, this effect being more obvious at lower cooling rates (range from 3 to 6°C/s). It should be noted that at these slow cooling rates the pearlite microstructure appears quite heterogeneous, above all in the case of recrystallized austenite, with the presence of relatively coarse ferrite units, well over the mean size.

4. Discussion

4.1. Austenite Conditioning

The results in Fig. 1 and Table 3 show that the finest austenite microstructure prior to transformation is achieved with steel 11VN for both deformation schedules. The presence of a finer initial austenite grain size can be considered as one of the main factors contributing to this refinement. As vanadium content increased from 0.05 to 0.10% (steels 5V and 10V), a relative finer recrystallized grain size was obtained after schedule A (from 81 to 60 mm), which also affected the SV value after schedule B (from 52 to 62 mm), although in both cases a similar deformation was accumulated during the final three passes. This refinement in the austenite is more relevant when, together with vanadium, nitrogen content was also increased (steel 11VN).

The higher vanadium content and the nitrogen present in 11VN steel make vanadium nitrides more effective in controlling grain growth during reheating. It has been reported previously that this can be due to a higher fraction and retarded dissolution of precipitates and also to a possible vanadium solute drag effect. As a consequence, the SV value increased from 62 mm (steel 10V) to 95 mm (steel 11VN).

4.2. Pearlite Microstructure

For a given cooling condition, Fig. 4 illustrates that the room temperature microstructure obtained from an unrecrystallized austenite has a lower hardness values in the three microalloyed steels, while Fig. 6 suggests that this effect can be related to an increase in the interlamellar spacing of the pearlite. This increase in the interlamellar spac-
ing, between recrystallized and unrecrystallized microstructures, could be related to an acceleration in pearlite transformation (starting at higher temperatures and shorter times) due to the increase in the specific boundary area $S_V$ produced by the accumulation of strain (larger number of available nucleation sites), in a similar way to what occurs with ferrite transformation in low carbon Nb microalloyed steels when controlled rolling is applied.\textsuperscript{18–20)}

In all cases, an increase in the cooling rate led to a refinement of the interlamellar spacing and to an increase in the hardness. This is the expected behavior since increasing the cooling rate means that the undercooling increases and therefore the interlamellar spacing decreases. Independent of this general behavior, some differences were observed between the three steels. Comparing the microstructures obtained following schedule A, the relative small change in $S_V$ between steels 5V and 10V (25 and 33 mm\(^2\)/H\(_{11002}\), respectively) produced similar interlamellar spacing values for a given cooling rate (Fig. 6). However, an additional increase in $S_V$ obtained with steel 11VN (54 mm\(^2\)/H\(_{11002}\)), which in agreement with the aforementioned behavior should have increased the interlamellar spacing, resulted in the refinement of $\lambda$ at 3°C/s when compared to 5V and 10V steels, and the formation of M-B constituents at 6 and 8°C/s cooling conditions (Fig. 5). These results confirm an increase in the hardenability in steel 11VN, suggesting that the larger contents of Mn and N present in the chemical composition of this steel may be responsible for this increase.\textsuperscript{21) It is worth emphasizing that Staiger et al. observed a similar effect when nitrogen content increased from 30 to 70 ppm in the formation of hard constituents in the case of low carbon wire-rods.\textsuperscript{22}) In contrast, in agreement with the results shown in Fig. 5, steel 11VN has lower hardenability after schedule B, mainly due to the effect of accumulated strain.

4.3. Effect of Process Parameters on Hardness

The hardness of the pearlitic microstructure can be calculated from the sum of the following contributions:

$$HV = HV_{\text{sol}} + HV_{\text{prec}} + k \cdot \lambda^{-0.5} \quad \text{(4)}$$

where $HV_{\text{sol}}$ and $HV_{\text{prec}}$ represent the contribution of solid solution elements and vanadium precipitates in ferrite, respectively. The other factor is the effect of the interlamellar spacing of pearlite.

From the previous sections it can be concluded that in the terms that include the interlamellar spacing and the precipitation hardening, different factors are intervening simultaneously. Some of them can be summarized as follows:

- For a given cooling rate the austenite state, recrystallized or pancake, affects the interlamellar spacing.
- An increase in N and V contents favors an increase in

Fig. 8. “Orientation unit” size distributions derived for the different steels and deformation-cooling conditions studied.
precipitation hardening but simultaneously reduces the contribution of interlamellar spacing when pancake austenites are produced (larger λ values).

In order to gather all these different interactions, a quantitative approach is required. This was achieved by taking the value from the study made by Clarke and McIvor as a starting point. These authors quantified the contribution of the different microstructural and compositional factors on the tensile strength of heat treated ultra-high carbon steels and proposed the following expression:

$$\sigma_f (\text{MPa}) = f_{\text{treat}} + 1029(\%C) + 152(\%\text{Si}) + 210(\%\text{Mn}) + 235(\%\text{Cr}) + 442(\%\text{P})^{0.5} + 5244(\%\text{N}_{\text{free}})$$ ..............................(5)

where the term “$f_{\text{treat}}$” includes the contribution of cooling rate and austenite grain size prior to transformation (both parameters defining the conditions of the heat treatment applied to obtain the pearlite microstructure).

In the current study the term “$f_{\text{treat}}$” was modified as follows:

- The effect of the austenite microstructure prior to transformation was considered by means of the specific grain boundary area, $S_y$, instead of the mean grain size. This means the application of Eq. (5) can be extended to pearlite microstructures transformed from pancake austenites following thermomechanical processes.

- The contribution of vanadium precipitation was included using the approach proposed by Lagneborg et al.23 for medium carbon vanadium microalloyed forging steels (with pearlite fractions ranging from 60 to 95%). The authors concluded that the contribution of precipitation to strength depends on the $N$ and $V$ contents and on the cooling rate ($CR$ in °C/s) in the following form:

$$\sigma_{\text{prec}} (\text{MPa}) \propto (5(\%N) + (\%V)) \cdot ((2.2 + 0.7 \log(CR)) \cdot f_{\text{treat}}) \cdot (2.2 + 0.7 \log(CR)) \cdot S_y \text{ (mm)}$$ ..............................(6)

That is, they observed that, from the point of view of precipitation strengthening, 5 parts per weight of $N$ are equivalent to one part of $V$.

When strain induced precipitation in austenite occurs, a part of the total nitrogen and vanadium is lost for further precipitation and strengthening of the steel during subsequent cooling after hot rolling. As a result, it is necessary to determine the amount of VN precipitated in austenite during rolling in order to properly evaluate the strengthening contribution of $V$. Based on the analysis done in Ref. 13) the first hypothesis was that 30% of the total possible equilibrium fraction was required to stop recrystallization and accumulate strain when the $T_m$ temperature was reached. Applying this approach to the tests corresponding to schedule B, it was estimated that less than 10% of the total nominal vanadium would be precipitated in austenite for the three steels. Therefore, it can be assumed that there will not be any significant differences in $V$ precipitation hardening between recrystallized and unrecrystallized microstructures and for simplicity, to evaluate the contribution of precipitation in Eq. (6) the nominal contents of $N$ and $V$ were taken into account (it is worth emphasizing that based on Thermo-calc analysis it can be concluded that with the selected soaking temperature all the vanadium and nitrogen are in solution before applying the deformation schedules).

The different microstructures were characterized by hardness measurements and the following relationship between the tensile strength and Vickers hardness values, obtained for pearlite steels microalloyed with vanadium, was applied45:

$$HV = 0.31 \sigma_f \text{.........................................}(7)$$

Taking into account Eqs. (5)–(7) and the experimental values from Fig. 4, Table 3, some previous results obtained for the plain C–Mn2 steel25 in Table 1 and those obtained with steel 11VN in Ref. 13), the following expression was obtained for the $f_{\text{treat}}$ term of Eq. (5):

$$f_{\text{treat}} = 262 \cdot \log(CR) - 154 \cdot \log(S_y) + 607 \cdot (5N + V) \cdot (2.2 + 0.7 \log(CR)) \cdot S_y \text{ (mm)}$$ ..............................(8)

where $CR$ is in °C/s and $S_y$ in mm$^{-1}$.

It is worth emphasizing that Eq. (8) confirms the reduction in strength that is associated with a refinement in the austenite microstructure prior to transformation, quantified as an increase in $S_y$. The beneficial effect of increasing the cooling rate, that is, a reduction in the interlamellar spacing and a promotion of a finer and more copious $V(C, N)$ precipitation, is also well described in the equation.

The comparison between the predictions of Eq. (5) (with $f_{\text{treat}}$ from Eq. (8)) and the experimental hardness values is illustrated in Fig. 9. It should be noted that the results corresponding to schedule B do not deviate significantly from the predictions obtained with microstructures transformed from recrystallized austenite, confirming that the possible loss of precipitation hardening as a consequence of an early precipitation of VN in austenite during deformation, is small.

In order to check the validity of the above equations on a wider range of steel compositions and cooling conditions the data reported by Pickering and Garbaz26 and Jaiswal and McIvor5 for eutectoid steels including plain carbon and vanadium microalloyed steels, were considered. The results correspond to cooling rates from 5 to 35°C/s and values of the term (5N + V) between 0.065 and 0.24. The predictions of Eqs. (5) and (8) for these data are shown in Fig. 10.
From the figure it can be observed that the predictions are fairly good for the plain carbon steel at all the cooling rates considered. However, for the vanadium steels there is a tendency to overestimate the hardness for the data corresponding to cooling rates higher than 10°C/s, although at the lowest cooling rate of 5°C/s the predictions are reasonable. This behavior suggests that the contribution of vanadium precipitation in Eq. (8), which depends on vanadium and nitrogen content as well as on the cooling rate, is being overestimated for high cooling rates. A new regression analysis taking into account the data in Refs. 5) and 24) produced the following equation that can be applied at cooling rates equal to or higher than 10°C/s:

\[ f_{\text{treat}} = 262 \cdot \log(CR) - 154 \cdot \log(S_f) + 34 \cdot (5N + V) \cdot (59 - 20 \log(CR)) \]  

(9)

From Eq. (9) it should be noted that in this range of cooling rates, an increase in the cooling rate led to a reduction in vanadium precipitation hardening, in contrast to what occurs at lower cooling rates (<6°C/s) where vanadium precipitation is enhanced when the cooling rate was increased. Figure 11 shows the good predictions obtained for all the materials using two equations, Eq. (8) for cooling rates \( \leq 8^\circ\text{C/s} \) and Eq. (9) for cooling rates \( \geq 10^\circ\text{C/s} \), respectively.

Finally, it should be mentioned that a limitation in the vanadium precipitation hardening effect is clearly observed for the pearlitic steels analyzed in this study. This behavior was also seen in the data taken from the literature,\(^{5,24,25}\) with a maximum hardness value of about 400HV which could not be increased by further increasing the vanadium content. It has been suggested that this limiting precipitation hardening might be a consequence of the limited amount of C and N dissolved in the ferrite and available to form precipitates.

4.4. Crystallographic Ferrite Unit

In Fig. 12, the mean values of “orientation unit” size, \( D_p \), were plotted as a function of the austenite grain boundary area \( (S_V) \) for different cooling rates. Figure 11 shows the good predictions obtained for all the materials using two equations, Eq. (8) for cooling rates \( \leq 8^\circ\text{C/s} \) and Eq. (9) for cooling rates \( \geq 10^\circ\text{C/s} \), respectively.

Finally, it should be mentioned that a limitation in the vanadium precipitation hardening effect is clearly observed for the pearlitic steels analyzed in this study. This behavior was also seen in the data taken from the literature,\(^{5,24,25}\)
with $D_p$ in $\mu m$, $CR$ in ($^\circ C/s$) and $S_p$ in mm$^{-1}$. Figure 13 shows the correlation between experimental and calculated values of the “ferrite unit” size given by Eq. (10). It must be pointed out that the application of this equation is limited to the range of cooling rates that it was derived in (between 3 and 8$^\circ C/s$).

The relationship between the austenite and pearlite microstructure can be explained by the model proposed by Smith,27 who considers the orientation of ferrite in pearlite to be dependent on the neighboring grain of austenite. In this case the relationship between pearlite colonies is determined by the neighboring austenite grain, and similarly oriented colonies form an orientation unit which becomes a single cleavage facet after fracture. As the grain size decreases, the size of the orientation units decreases. The smaller the prior austenite grain size (the higher the $S_p$), the higher the number of orientation units and therefore the higher the resistance to crack propagation. In this study the colony size was not measured, but it has been suggested that this is independent of the prior austenite grain size and depends only on the transformation temperature in a manner similar to the interlamellar spacing.8,9 Consequently, the decrease in the “orientation unit” size as the prior austenite grain size reduces ($S_p$ increases) would imply that these units should be formed by a decreasing number of colonies as it was observed by Garbaz et al.9

Equations (8) and (9) show that $S_p$ affects the strength (through the interlamellar spacing). On the other hand, Eq. (10) indicates that for a given cooling rate $S_p$ modifies the crystallographic unit size, $D_p$. Although in this study toughness measurements have been not performed, a wide range of previous published works confirm that a reduction in $D_p$ leads to an improvement in toughness.6,7,28,29 In contrast to what happens with low carbon steels, where ferrite refinement improves both strength and toughness, an increase in toughness through a refinement of $D_p$ will be accompanied by a decrease on the strength of pearlitic steels. This is illustrated in Fig. 14, where the evolution of $\sigma_T$ and $D_p$ as a function of $S_p$ were drawn for a specific base chemical composition (0.8C, 0.80Mn, 0.25Si, 0.30Cr) and a cooling rate of 6°C/s calculated with the help of Eqs. (5), (8) and (10). In the figure, the contribution of vanadium precipitation is included for different 5N+V contents. As mentioned above in the case of pearlitic steels with chemical compositions similar to that considered in Fig. 14, there is saturation in the maximum attainable tensile strength value that cannot be increased by a further increase in the V content. Taking into account the saturation values published in the bibliography,2,24,25 an interval has been drawn in the figure (shaded area).

For a given plain C–Mn pearlitic steel, an increase in toughness through decreasing $D_p$ parameter can lead to an important loss in tensile strength, as shown in Fig. 14. This reduction in strength can be compensated through precipitation hardening by the addition of V, which combined with a proper thermomechanical processing design, will also allow to lower $D_p$ values to be obtained. In these conditions, an optimum combination of nitrogen and vanadium content is very important to achieve good strength–toughness combinations.

Finally, it is worth emphasizing that, in addition to the reduction in the $D_p$ mean value for a given cooling rate, the pearlite transformation from an unrecrystallized austenite significantly narrows the size distribution of the “ferrite units” as shown in the results of Fig. 8, this effect being more noticeable at the lowest cooling rate. As fracture toughness in the brittle and ductile–brittle regime is controlled by a “weakest link” behavior, the reduction in the fraction of coarse $D_p$ units will improve toughness behavior by decreasing its dispersion, as suggested in Ref. 30).

5. Conclusions

(1) The application of thermomechanical treatments to V microalloyed eutectoid steels brings about a refinement of the “ferrite unit” microstructural feature controlling cleavage fracture. This refinement is more pronounced as vanadium and/or nitrogen contents are increased.

(2) In contrast to what occurs with low carbon steels, the “ferrite unit” refinement is also associated with a decrease in the strength due to an increase in the pearlite interlamellar spacing for a given cooling rate. This loss in strength can be compensated by optimizing V–N microadditions. The small quantity of VN required to be precipitated in austenite for strain accumulation during thermomechanical processing guarantees sufficient amount of free V available for further precipitation during transformation.

(3) An equation which relates the mean “ferrite unit” size of pearlite to $S_p$ and the cooling rate during transformation was derived.

(4) A new equation to predict pearlite strength as a function of $S_p$, cooling rate and V and N contents was determined.

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