Beitrag zur Untersuchung von niedrig gekohlten, mit Niob und/oder Vanadin mikrolegierten Schmiedestählen

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Ein Beitrag zur Untersuchung von niedriggekohlten, mit Niob und/oder Vanadin mikrolegierten Schmiedestählen

de Monsieur Hans Leber

UNIVERSITÉ DE NEUCHATEL

FACULTÉ DES SCIENCES

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THE COMBINED EFFECT OF NIOBIUM AND VANADIUM IN A FORGED LOW CARBON STEEL

by

H. Leber, H. Lüthy and W. Form

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Thèse présentée à la Faculté des sciences de l'Université de Neuchâtel

par

Hans Leber

Métallurgiste diplômé
de l'Université de Neuchâtel

pour l'obtention du titre de
Docteur ès sciences.

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THE COMBINED EFFECT OF NIOBIUM AND VANADIUM IN A FORGED LOW CARBON STEEL

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Abstract

A previously chemically optimized low carbon Nb/V forging steel was subjected to a series of specific thermomechanical treatments. This was done in order to better understand the individual and combined effects of the microelements in question on the interplay between dissolution of their precipitates during austenitizing and reprecipitation in the austenite as well as during or after subsequent transformation into either ferrite and pearlite or bainite. Furthermore, the influence of isothermal aging after hot forging and rapid cooling to 500-600°C on strengthening has also been examined.

As far as the latter aspect is concerned our results show that the increase in strength by aging the Nb/V steel is raised beyond the sum of the individual effects produced by each microelement alone, provided a bainite matrix is attained.

Introduction

In an earlier study (1) - one of the first that dealt with a low carbon microalloyed forging steel - it was shown that in a Nb/V steel a beneficial strengthening effect, with excellent impact properties could be obtained with thicknesses up to at least 120 mm through controlled thermomechanical processing. This opened up a new field with considerable potential for the forging industry which hitherto had essentially confined its interest to medium carbon microalloyed steels. During this earlier study the chemical composition of a series of low carbon microalloyed steels was optimized relative to strength and impact properties through the method of multifactorial planning. The optimized steel was then subjected to a number of thermomechanical treatments which were conducted in that temperature range in which austenitizing and subsequent forging is industrially carried out, followed by either air, oil or water cooling.

In the present work we examined more closely the interactions between the various phenomena known to occur in thermomechanically treated microalloyed steels in order to make a contribution to the understanding of the relationship existing between induced microstructures and resultant mechanical behaviour. These phenomena include:

- dissolution of the microalloy compounds precipitated during prior processing
- deformation and recrystallization (or absence of it) of the austenite
- decomposition of the austenite into either ferrite and pearlite or into bainite
- reprecipitation of the microalloys in the austenite on one hand, and in the ferrite or bainite, on the other hand.

Whereas the assessment of recrystallization and polymorphic transformation does not necessarily require sophisticated and tedious means of investigation, the analysis of precipitates, notably as far as coherent particles are concerned, is of considerable difficulty, and has in the present work mostly been done by more or less direct inferences, associated with a considerable degree of speculation. For instance, throughout this study, and based on literature references, we have assumed that strengthening due to precipitation is due, above all, to carbonitrides of either niobium or vanadium.

In our aim to contribute to the understanding of microstructures resulting from thermomechanical forging we have concentrated on the following aspects:

- effect of solutioning parameters on metallurgical phenomena taking place during subsequent thermomechanical processing. This was a preliminary study on which we shall not report extensively here;
- effect of forging temperature and holding time after forging on the final microstructure and on strength;
- effect of aging between 500 and 600°C following rapid quenching from the forging to the aging temperature on additional strengthening.

In addition to the work carried out on the optimized Nb/V steel, experiments were also conducted on a steel containing only Nb and on another one containing only V so as to permit to ascertain separately the pertinent effects of these 2 microelements.

Materials

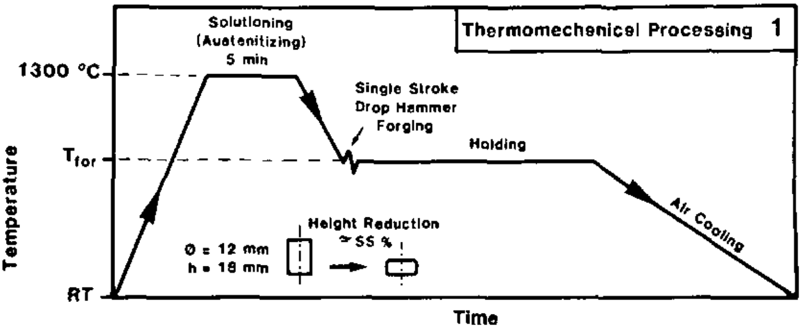
Table I lists the chemical composition of the microalloyed steels studied, the first one listed being the chemically optimized Nb/V steel, which is our major study material. A reference steel, having the same basic composition as the other 3 but no microelements has been added for sporadic comparison. All 4 steels were available in the form of forged bars.

Table I: Chemical composition in wt % of the investigated steels

Steel	C	Mn	Si	S	P	Al	Nb	V	N
Nb/V	0.12	1.45	0.35	0.014	0.009	0.036	0.07	0.07	0.0115
4Nb	0.09	1.45	0.35	0.005	0.011	0.027	0.065	(0.008)	0.0068
2 V	0.11	1.39	0.30	0.004	0.011	0.027	(0.002)	0.089	0.0068
R	0.12	1.47	0.35	0.003	0.008	0.029	-	(0.008)	0.0080

Procedure

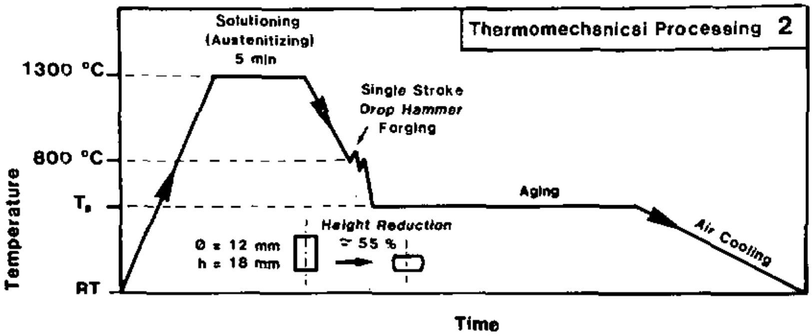
The main part of our study consisted in carrying out 2 types of thermomechanical processings. In processing 1, depicted in Fig. 1, forging temperature and holding time at this temperature following forging were varied. Processing 2 is illustrated in Fig. 2, its principal variables being time and temperature of aging following rapid cooling from the forging temperature. For both processes a soaking temperature in the austenite of 1300°C was used, since at this temperature all



Forging Temperature T_{for} = 700, 800, 900, 1000 °C

Holding Time at T_{for} = 0, 0.12, 1.2, 12, 40, 120 min

Fig. 1 - Schematic temperature-time diagram of the thermomechanical processing 1.



Salt Bath Temperature T_s = 500, 550, 600 °C

Aging Time at T_s = 0.12, 1.2, 12, 40, 120, 400, 1200 min

Fig. 2 - Schematic temperature-time diagram of the thermomechanical processing 2.

Nb - and therefore also all V - should be in solution in our low carbon steel (2), guaranteeing maximum precipitation hardening potential. The austenitizing/solutioning treatments were done in a high temperature muffle furnace. The thermomechanical treatments were carried out on small cylindrical specimens, 18 mm high and 12 mm in diameter, machined from the forged stock available from our previous study (1). After treatment according to either processing 1 or 2, specimens were machined or cut for mechanical testing, metallography and hardness measurements.

Forging was effected on a single stroke drop hammer reducing the height of the cylindrical specimens by 55% at a strain rate estimated to be 240 s^{-1}. In process 2 the samples, after forging, were immediately quenched into a salt bath, held between 500 and 600°C, for aging purposes.

In view of the small final dimensions of the forged material only a compression test specimen in the form of a tiny cylinder of 6 mm height and 3 mm diameter could be machined from its center portion. On this compressed specimen $Rp_{0.2}$ was measured in an universal tensile/compression testing machine. Thereby it was assumed that its value corresponds approximately to that one would have measured in tension, which has been verified in a series of comparative tensile tests. The tensile strength was calculated from hardness measurements using the relation: R_m [MPa] = 3.1 x HV 30 [kp/mm^2] (1).

The microstructure of the austenite revealed after quenching and slight tempering, the final microstructures obtained through processing 1 and 2, and those at various stages of aging were mainly observed in the optical microscope using standard procedures. Grain size determinations of either austenite or ferrite were made by the well known intercept method (3).

For unequivocal identification of bainite, observations were also made in the SEM, were deemed necessary.

Finally, X-ray identification of isolated precipitates from a few selected samples were carried out at the laboratories of Sulzer Brothers in Winterthur (Switzerland).

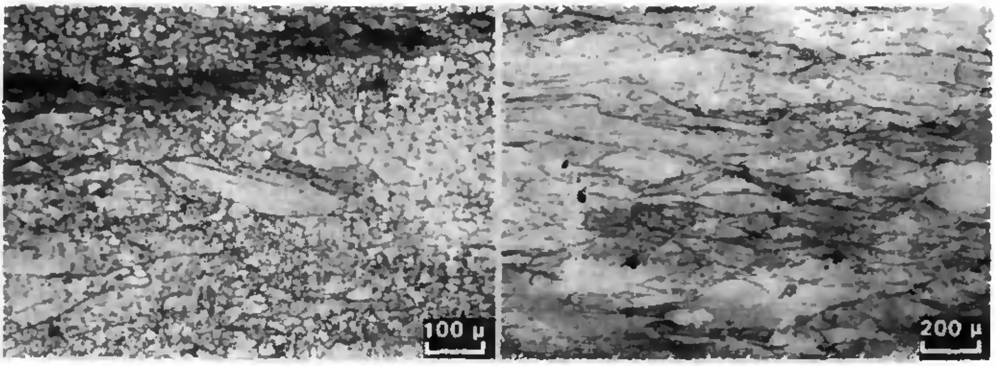
Results and Discussion

Influence of Austenitizing Time and Temperature

This preliminary study was only carried out with the Nb/V steel. We first selected an austenitizing temperature of 1300°C for the reasons given previously assuming that this temperature probably lies above the grain coarsening temperature for this steel (4). Indeed, substantial grain growth sets in within 2 min. producing an average grain size of about 400 μm. Further holding for up to 20 min. does not substantially change this size. A soaking time of 5 min. in the muffle furnace therefore seemed to us appropriate for the thermomechanical processing 1 and 2.

Immediately after forging at 1000°C a partly recrystallized austenite is observed (Fig. 3a), whereas immediately after forging at 800°C no recrystallization can be seen as is evident from the typical pancake morphology of the grains (Fig. 3b). After air cooling to room temperature both microstructures consist of ferrite and pearlite and of bainite, the samples forged at 800°C having a somewhat finer ferrite grain.

When auatenitizing at 1050°C, where less than 25% of the total Nb content is dissolved, no coarsening is observed after 2 min., and further soaking for up to 20 min. increases grain size only slightly from an initial 23 μm to 28μm. Immediately after forging at 1000°C complete recrystallization is reached with an average austenite



a) forged at 1000°C and immediately water-quenched.

b) forged at 800°C and immediately water-quenched.

Fig. 3 - Microstructures of the Nb/V steel, austenitized at 1300°C

grainsize of 18µm. Again, after forging at 800°C no recrystallization is visible, but some ferrite formation has been induced through deformation. After air cooling to room temperature the microstructure is completely ferrite-pearlite for both forging temperatures. These observations may be explained in the following manner:

- precipitated particles persisting at 1050°C do not coalesce at this temperature, thus preventing austenite grain coarsening;
- the higher resistance to recrystallization at 1000°C after austenitizing at 1300°C when compared to 1050°C may readily be explained by the more extensive Nb solubility at the higher temperature. It is well known (5) that Nb in solution retards recrystallization. In addition, the very fine grain after soaking at 1050°C contains many more sites of potential recrystallization nuclei than does the much larger grain size obtained at 1300°C.
- Nb in solution lowers the transformation temperature to ferrite (6). This explains the higher amount of ferrite in the samples solution-treated at only 1050°C.

Effect of Forging Temperature and Holding Time (Processing 1)

This part of our study was conducted on the three microalloyed steels listed in Table I. The microstructure and mechanical properties obtained after final air cooling to room temperature are summarized in Table III. The holding time of 40 min. was the longest one applied to the 4 Nb steel, whereas for the other two steels holding was extended to 120 min., which did, however, not substantially change the results obtained after 40 min.

The three steels had the following average austenite grain size after austenitizing at 1300°C for 5 min. (Table II):

Table II: Austenite Grain Size of the 3 Microalloyed Steels after Austeniting at 1300°C for 5 min.

Steel	γ-grain size [µm]
Nb/V	390 ± 40
4 Nb	194 ± 7
2 V	280 ± 33

It is interesting to note that adding V to Nb reduces the resistance to grain coarsening more than one would expect from the effect of V alone. This result confirms previous observations (7). Before we look in more detail at Table III let us briefly summarize the effect of forging temperature and holding time on the recrystallization kinetics. This is represented in the time-temperature-recrystallization diagrama of Fig. 4. As we have already seen from our preliminary experiments discussed in the preceding paragraph, there is no recrystallization in the Nb/V steel immediately after forging at 800°C, and only traces can be detected after holding for more than 12 min. On the other hand, recrystallization sets in right after forging at 900°C and reaches about 40% after 2 hrs. The 4 Nb steel recrystallizes already at 800°C, but reaches less than 15% after 40 min. It is interesting to note for this steel that, although recrystallization sets in at 900°C immediately upon completion of forging, only about 50% of the deformed regions have recrystallized after 40 min.

Finally, in the 2V steel recrystallization sets in immediately on forging at 800°C and reaches 100% within a few seconds. Thus, while it is known (8) that V retards recrystallization only slightly and that Nb in solution does so to a markable degree, our results show that the combined effect of Nb and V is even more pronounced that one would expect from the individual effects of the microelements. These recrystallization tendencies may also be sorted out from the results in Table III, although in a less complete manner.

As far as the amount of transformed matrix is concerned we have made a more complete study on the Nb/V steel than on the two others. This permitted us to trace the TTT-curve for the austenite to ferrite transformation illustrated in Fig. 5. Here we see that after forging at 900°C the ferrite transformation sets in only after 12 min., whereas after forging at 800°C ferrite formation begins immediately, but - and this seems to be an interesting aspect - proceeds extremely slowly. Thus, even after holding for 2 hours, no more than about 5% ferrite have formed. This is in contrast to ferrite formation after forging at 700°C, which sets in also immediately, (less than 2 sec.) but reaches 75% within 2 hrs. Taking into account that the Ar_3 temperature for this steel without deformation lies at 738°C (9) and that the delay time at this temperature is about 2 min. (7), our results confirm that forging not only increases Ar_3 considerably (from 738°C to about 900°C), but may also reduce the delay time by a factor of more than 50.

Returning to Table III it is interesting to note that the mean ferrite grain size is for all conditions and all microalloying steels the same and lies around 10 μm. Thus the large differences in initial austenite grain size has been completely wiped out through forging. Since it is known (10) that Nb in solution favors formation of bainite, we have also determined the amount of this phase after completion of processing 1, i.e. after final air cooling to room temperature. As expected, the 2V steel does not contain any bainite. Surprisingly enough the 4 Nb steel contains only a small amount of it when cooled right after forging at either 800 or 900°C, and none when held for 40 min. before air cooling. Thus, the 4Nb and the 2 V steels have a ferrite-pearlite matrix. On the other hand, the combined effect of Nb and V on bainite formation is very strong.

Without holding after forging the resultant microstructure contains between 25 and 55% bainite, practically independent on forging temperature. For this steel, too, the amount of bainite seems to become somewhat smaller with increasing holding time, notably at the lower forging temperatures.

As far as tensile test results are concerned (determined from compression and hardness test respectively) the last four columns in Table III give the values for $Rp_{0.2}$ and for R_m. There is a continuous decrease in both strength levels with increasing holding time. On the other hand, lowering the forging temperature increases the strength level. In Fig. 6 we plotted the yield strength of the three microalloyed steels

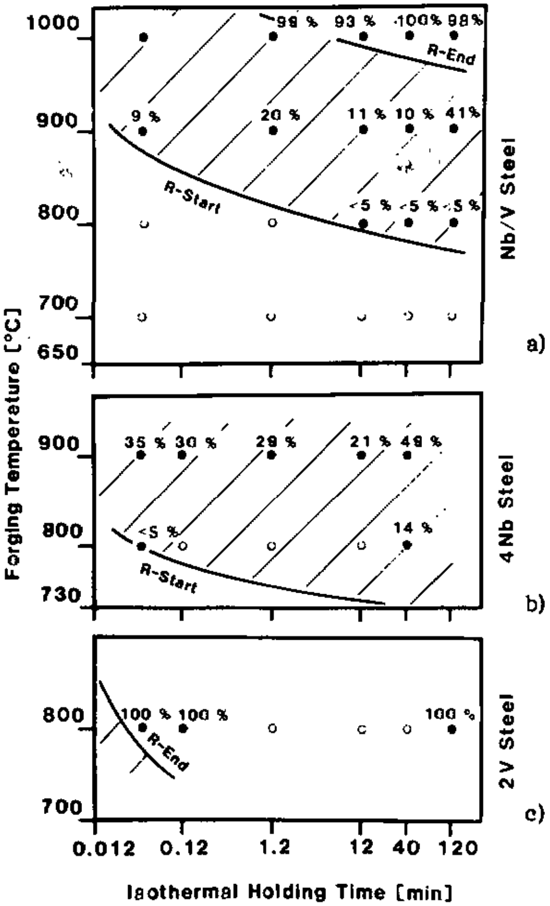


Fig. 4 - Time-temperature-recrystallization diagrams for the forged austenite (austenitizing temperature = 1300°C). The numbers indicate the percent recrystallized area.

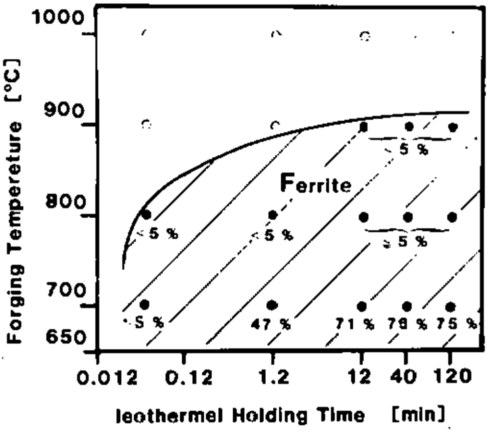


Fig. 5 - Time-temperature-transformation diagram of the Nb/V steel, held after forging at the time and temperatures indicated.
The numbers reflect the percent area transformed to ferrite.

Table III: Effect of forging temperature and holding time during processing I on final microstructure and mechanical properties of the 3 microalloyed steels.

Steel	T_{for} [°C]	Microstructure Recrystallizat. of γ [%] after		Bainite [%] besides F + P after t_H[min] and air cooling		Ferrite grain size [µm]	Mechanical properties after t_H[min] and air cooling $R_{p0.2}$ [MPa]		R_m [MPa]	
		t_H = 0 min	t_H = 40 min	t_H = 0 min	t_H = 40 min		t_H=0	t_H=40	t_H=0	t_H=40
Nb/V	1000	93 %	→ 100 %	25 - 55 %	25 - 40 %	12 - 15	534	466	737	617
	900	9 %	20 - 40 %	35 - 45 %	20 - 35 %	9.1 - 10.3	554	473	729	595
	800	0 %	< 5 %	40 - 55 %	15 - 35 %	9.4 - 10.5	579	525	747	629
	700	0 %	0 %	30 - 50 %	15 - 25 %	4.3 - 13.5	604	501	768	629
4 Nb	900	30 %	40 %	~ 5 %	< 5 %	10.5 - 11	514	400	710	560
	800	< 5 %	14 %	~ 5 %	< 5 %	9 - 9.4	548	460	692	584
2V	800	100 %	100 %	< 5 %	< 5 %	9.4 - 10.4	484	445	577	546

T_{for} = Forging temperature
t_H = holding time at T_{for}
F = Ferrite, P = Pearlite

as a function of holding time after forging at 800°C, to better illustrate the general tendencies. The lines drawn between the measured points reflect the average linear decrease of $Rp_{0.2}$. Not surprisingly, the Nb/V steel has the highest yield strength, followed by the 4 Nb and the 2 V steels. This trend is also typical for the tensile strength, irrespective of forging temperature.

If our Nb/V steel had an entire ferrite-pearlite microstructure we could calculate the contribution of precipitation to $Rp_{0.2}$ through the equation of Gladman et al. (11). However, the bainite portion of the matrix is too high to do this. Therefore, comparison of precipitation hardening contributions in the 3 microalloyed steels can only be done on the basis of the measured $Rp_{0.2}$ values, which, then, is affected in an unknown manner by differing matrices.

According to Gladman et al. (11) a ferritic-pearlitic carbon steel with an average grain size of 10µm containing 1.45% Mn and 0.35% Si, but no microelements should have a yield strength of 322 MPa. Compression tests carried out on our R steel of similar basic composition and grain size gave a value of 355 MPa for $Rp_{0.2}$, which is in good agreement with the calculated value. We assume, then, that any value above 355 MPa reflects precipitation strengthening occurred during processing 1.

Looking at the results in Fig. 6 we must conclude that in the Nb/V steel precipitation during final air cooling was most pronounced. Two reasons may be considered for this: first, this steel contains twice as much of microelements (Nb and V together) than the other two steels and therefore may form more precipitates. Second, V in solution retards Nb precipitation in the austenite and this apparently more than offsets the accelerating effect produced by forging. Therefore, without holding at the forging temperature precipitation will really proceed only once austenite has transformed. The latter reason seems to us to be more probable, since V cannot readily precipitate in the highly bainitic matrix of the Nb/V steel (12), so that its contribution to precipitation hardening (as inferred by the first reason) may be far below that expected from its actual content in the steel. On this basis of the above argument it follows also that in the absence of V more Nb may already precipitate in the austenite of the 4Nb steel than in the case of the Nb/V steel. Consequently, its precipitation strengthening potential in the transformed matrix is reduced compared to the Nb/V steel.

The continuous decrease in strength with holding time must reflect a continual precipitation of Nb (or of V in the 2V steel) in the austenite during this period, notably at the higher temperatures, on one hand, and progressive coagulation of the precipitates at all forging temperatures, on the other hand. This again reduces the precipitation strengthening potential. The fact that Nb precipitates in the austenite is supported by the observation that the amount of bainite in the transformed matrix decreases with increasing holding time. Since no grain coarsening occurred, no contribution from it to the observed softening tendency with time must be taken into account.

An interesting point concerns the 2V steel. Since its $R_{p0.2}$ value is higher than that of the R steel it follows that precipitation in the ferrite has also taken place despite the fact that our cooling rate in air ($V_{R}{}_{500}^{800}$ = 300°C/min.) is about six times higher than the maximum permissible cooling rate for V precipitation as determined by Charlier (12).

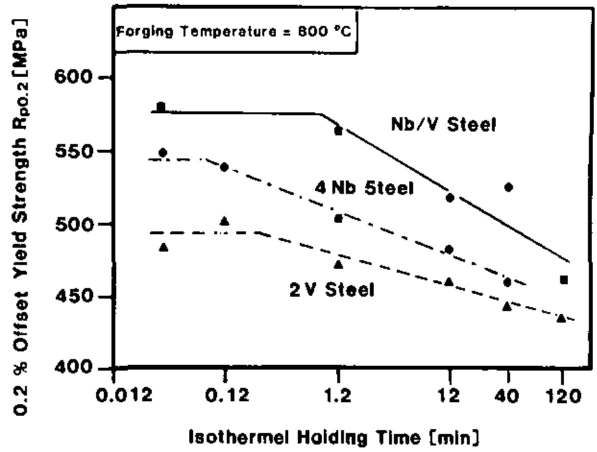


Fig. 6 - $Rp_{0.2}$-yield strength as a function of holding time after forging at 800°C for the 3 microalloyed steels.

In this connection we may also point out that the highest measured $Rp_{0.2}$-value (604 MPa after forging at 700°C and immediate cooling) is lower than that reached in a prior study (1) with the same Nb/V steel. In that study the air cooling rate was appreciably slower because of much larger specimens. It may then be that because of a too rapid cooling rate in the present study, the potential of V precipitation could not be properly exploited.

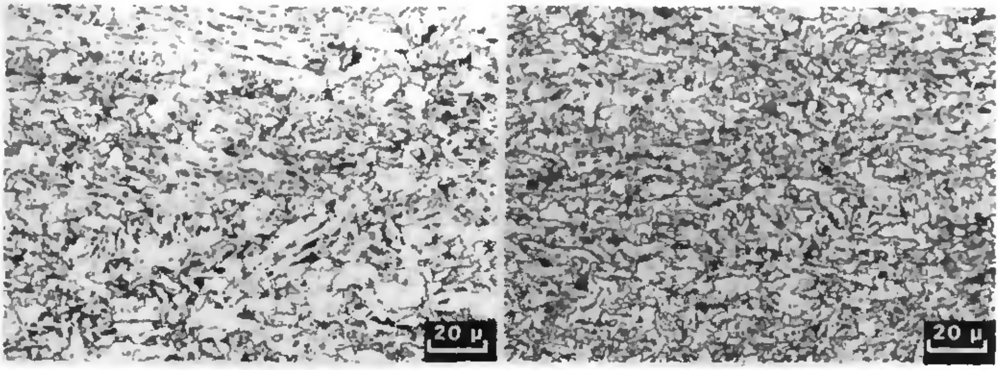
Effect of Aging Between 500 and 600°C Following Forging (Processing 2)

In applying the thermechanical processing 2 to our three microalloyed steels we intended to use conditions under which their age hardening potential ought to deploy itself to its fullest extend: First, they were austenitized at a temperature at which all microelements are in solution, then they were forged at 800°C, which is about the lowest industrially feasible temperature, whereupon they were immediately quenched into a salt bath at 500-600°C so as to make sure that at the start of aging the highest portion of microelements was still in solution.

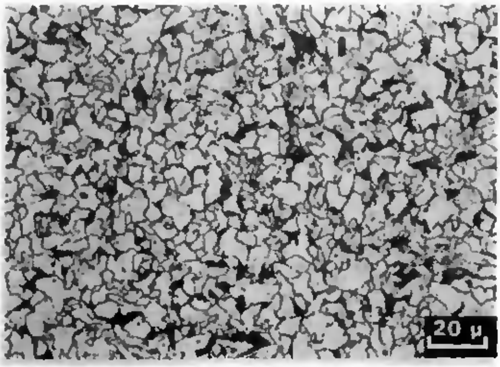
The microstructural observation and mechanical properties obtained during this part of our study are summarized in Table IV. Let us first comment on the microstructures developed after going through the entire processing 2. Assuming that a substantial amount of the microelements are still in solution when the samples reach the salt bath temperature we would expect in the two Nb containing steels a strong tendency to bainite formation. This is indeed the case es reported in Table IV.

Fig. 7 illustrates the microstructures in the three steels after aging at 550°C, which confirm our expectations. In eddition, an analysis of the evolution of the microstructures during aging shows that in the samples of the Nb/V steel aged at 600°C the amount of bainite diminishes with time.

This indicates that the austenite transformation is incomplete when the sample reaches this salt bath temperature so that the austenite becomes progressively depleted in Nb, and therefore tends more and more to transform to a ferrite-pearlite



a) Nb/V steel after aging for 1.2 min. b) 4Nb steel after aging for 1.2 min.



c) 2V steel after aging for 1.2 min.

Fig. 7 Microstructures of 3 microalloyed steels obtained after forging at 800°C, and quenching to the aging temperature of 550°C and subsequent air cooling to room temperature

matrix during air cooling to room temperature. This trend is even more pronounced with the 4Nb steel, in which Nb precipitates more rapidly in the absence of V. The 2 V steel matrix transforms to a virtually complete ferrite-pearlite structure at any time and temperature of aging. To be sure, in addition to changes in type and proportion of the transformation products, the microstructures become coarser with time and temperature. Also, the bainite formed at the beginning, becomes more and more equiaxed, and the pearlite globularizes.

X-ray identification of isolated particles from the Nb/V and the 4Nb steels indicates a much smaller yield than the one obtained on the same Nb/V steel in a prior study (1). This suggests that there were less incoherent particles in the present study, most likely because of a higher austenitizing temperature (1300°C compared to 1200°C). Furthermore, few incoherent particles are found when quenching in water or cooling in air directly from the 800°C forging temperature. It seems then as if cooling rate does not so much affect precipitation kinetics as it does the decomposition of austenite.

Table IV: Effect of time and temperature of aging following forging at 800°C on final microstructure and mechanical properties.

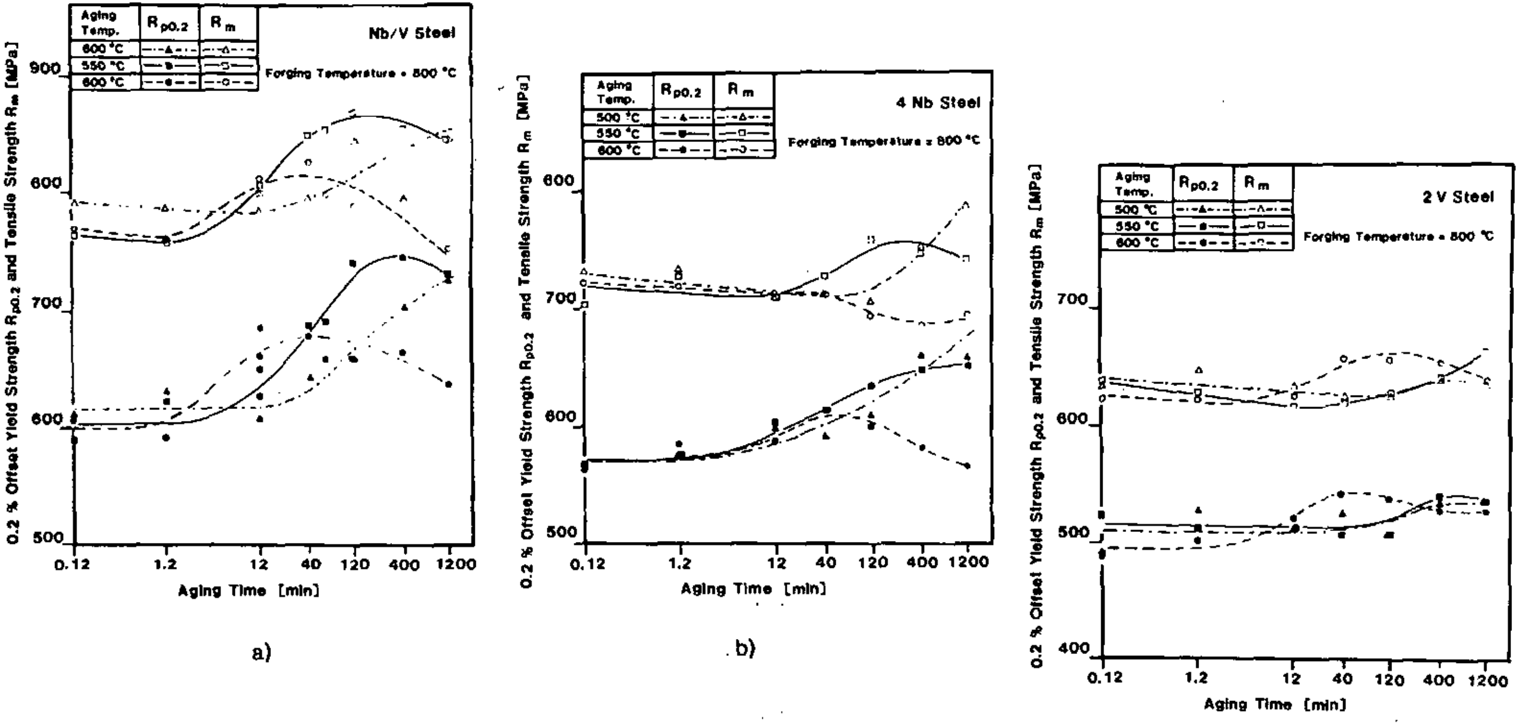
Steel	T_s [°C]	Microstructure observed after aging and cooling to room temperature	Mechanical properties at the start of aging and at the age-hardening maximum					
			$R_{p0.2}$ [MPa]			R_m [MPa]		
			Initial value	Max. value	after t_A [min]	Initial value	Max. value	after t_A [min]
Nb/V	500	Bainite	617	727*	1200	794	846*	1200
	550	Bainite	606	746	400	766	871	120
	600	Bainite; beyond 1.2 min.:polygonal Ferrite	600	679	40	772	828	40
4 Nb	500	Bainite	571	660*	1200	722	787*	1200
	550	Bainite + F(d_α = 3-5 µm) + P	572	651*	1200	716	759	120
	600	Bainite + F($d_\alpha \to$ 10 µm) + P + tertiary-Fe_3C	579	614	40	722	722	0.12
2 V	500	F (d_α = 5-7 µm) + P + bainite(<5 %)	511	536*	1200	640	639	400
	550	F (d_α = 5-7 µm) + P + bainite(<5 %)	515	541	400	639	663*	1200
	600	F (d_α = 6.5-8 µm) + P + bainite(<5 %)	504	543	40	623	657	120

T_S = Salt bath temperature

t_A = aging time at T_S

F = Ferrite, P = Pearlite, Fe_3C = Cementite

* = no maximum; highest value measured within 1200 min.



a)

b)

c)

Fig. 8 - Aging curves in terms of $Rp_{0.2}$ and R_m of the 3 microalloyed steels after forging at 800°C and quenching to the aging temperature.

The highest yield was obtained in samples aged during 20 hrs. Obviously these samples are strongly overaged. Strangely enough no V was detected in the residues. This implies that vanadium precipitates have remained coherent, which, in turn, means that no V has precipitated in the austenite, as already implied before. Finally, in those Nb/V steel samples which contain pearlite we find substantial amounts of Fe_3C, which phase is totally absent in full bainite samples, as anticipated. This is, for instance, the case in the sample which had been aged at 550°C for 20 hrs.

In Fig. 8 the yield and tensile strength values, as determined from compression and hardness tests, respectively, are plotted against aging time. The variation in strength for all 3 steels is typical for aging curves, i.e. the maximum reached is higher the lower the aging temperature, and the time necessary to reach it becomes longer. The initial and maximum values are listed in Table IV. As indicated by asteriks some "maximum" values are merely the values reached after the longest time examined. The real maximum value in these cases must thus be higher than that given in Table IV. Figure 9 permits a direct comparison of the $Rp_{0.2}$ -curves for the 3 microalloyed steels aged at 550°C. In the Nb/V steel $Rp_{0.2}$ increases from an initial value of 605 MPa after quenching to the aging temperature to 745 MPa after holding for about 2 hrs. To our knowledge this is the highest yield strength value ever recorded with a 0.12% C steel. At the same aging temperature the 4Nb steel starts with an $Rp_{0.2}$ of 570 MPa and obtains its maximum value of 650 MPa after 20 hrs. The 2V steel increases its yield strength through aging at 550°C from an initial 515 MPa to only 540 MPa.

A comparison between the maximum age hardening effect of the three steels after aging at 550°C and 600°C, respectively, is made in Fig. 10. One sees that at 550°C the combined effect of Nb and V leads to an increase in yield strength of 150 MPa, whereas the sum of the increases due to aging in the 4Nb and 2V steels only amounts to 105 MPa (79 MPa + 26 MPa). The superiority by 45 MPa of the Nb/V steel expresses a synergistic effect produced by the two microelements when they are present simultaneously. After aging at 600°C on the other hand, the increase of $Rp_{0.2}$ in the Nb/V steel is about the same as the sum of the individual increases in the 4Nb and the 2V steels.

The combined effect of the microelements in the Nb/V bainitic steel may be explained by the following 2 interactions which occur consecutively:

The first interaction occurs in the austenite immediately after hot working at 800°C. The latter should promote precipitation of both Nb and V. Yet the presence of dissolved V hinders the precipitation of Nb and vice versa. As a result of this mutual hindrance precipitation in the austenite is minimized and, consequently, the aging potential in the transformed matrix maximized.

Let us look at the second interaction, and this at the aging temperature of 550°C first. Some very fine precipitated particles may possibly form in during quenching to the salt bath temperature. Transformation to bainite - favored by dissolved Nb - practically prohibits precipitation of V. But as long as V remains in solution it retards precipitation of Nb, since it increases the solubility of this element both in the austenite and in the transformed matrix. Thus, the major contribution to a higher initial yield strength than that obtained with the microelement-free R steel must be due to a microstructural effect originating from a more or less pronounced amount of bainite when forging at 800°C and immediately quenching to the salt bath temperature. Only with prolonged holding time will Nb precipitate (followed by V) as very fine particles from the supersaturated acicular (bainitic) structure. The formation of bainite is therefore a necessary condition for maximizing the combined effect of 2 microelements which consists in an important additional strengthening during aging at 550°C.

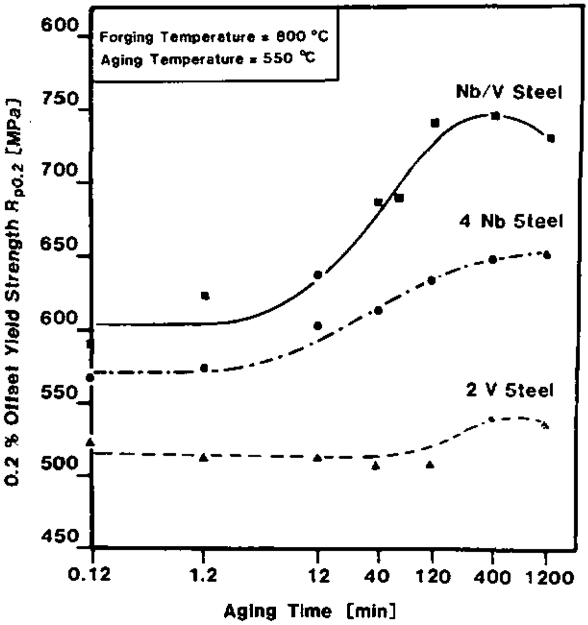


Fig. 9 - Aging curves in terms of $R_{p0.2}$ of the three microalloyed steels after forging at 800°C and quenching to an aging temperature of 550 °C.

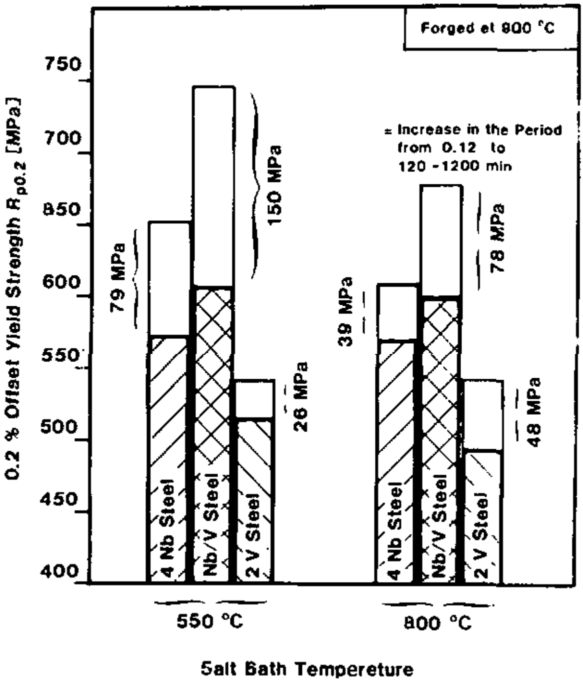


Fig. 10 - Comparison of the increases in yield strength from the start (shaded bars) to the maximum value measured for the 3 microalloyed steels after aging at 550°C and 600°C respectively.

At the aging temperature of 600°C, where the microstructure of the Nb/V steel is mainly ferrite-pearlite, the first interaction remains the same. The second one, however, is largely non operative in view of the small amount of bainite present at this temperature.

The above reasoning explains why the increase of strength by precipitation of microelements during aging is raised beyond the level expected from the individual effects of Nb and V when the steel is microalloyed simultaneously with Nb and V. This phenomenon is therefore most pronounced in a fully bainitic structure.

Summary

Most of the microstructural results obtained in this study on forging steel, can be qualitatively explained on the basis of effects of Nb and of V on recrystallization and on decomposition of austenite as well as on the age hardening potency, known from microalloyed coiled sheet steels.

Thus we confirmed the following tendencies:

- When Nb is fully dissolved rapid grain coarsening sets in during austenitization.
- In the presence of partially dissolved precipitation the matrix offers a much lower resistance to recrystallization following forging than a matrix in which such precipitates are fully dissolved during austenitization.
- The larger the amount of Nb in solution the stronger the tendency to bainite formation.
- The effect of V on recrystallization and on the tendency to bainite formation is negligible. However, addition of V to Nb increases strongly the resistance to recrystallization as well as the tendency to bainite formation.
- The continual decrease in strength with holding time after forging between 1000 and 700°C, may be readily explained on the basis of coarsening of non-coherent precipitates, and/or continuing recrystallization (above 800°C), the former effect being more pronounced as the results at 700°C confirm.

Additional strengthening, beyond the purely thermomechanical effects may be obtained through aging between 500-600°C following rapid quenching from the forging temperature. In fact, when aging for about 2 hrs. at 550°C an extraordinarily high yield strength value of 745 MPa is attained. This must be ascribed to the combined role played by Nb and V in strongly retarding precipitation in the austenite, on one hand - thus making available the full age hardening potential - and in favoring the bainitic transformation, on the other hand.

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