Fatigue of metallic components is an important aspect in mechanical engineering. For example, the repeated pressure differences result in the fatigue of the fuselage of aircrafts. These high demands require the application of endurable and lightweight materials. In this context, laminated metal composites produced by the accumulative roll bonding process can fulfill both aspects: outstanding mechanical strength of lightweight materials, due to an ultrafine-grained microstructure, combined with the possibility to produce fatigue-resistant materials by varying the stacking order of the laminated metal composites. This work investigates laminated metal composites consisting of aluminum AA2024 and titanium grade 1 layers with different stacking orders in respect to their three-point bending fatigue properties. Additional finite element method (FEM) calculations of the stress distributions in the laminated metal composites are used to explain the results. The mechanical properties of the laminates with an outer aluminum or titanium layer are discussed based on the investigations of nanohardness and crack propagations. Although maximum bending stresses at the sheet surface of laminates with outer Titanium layer are higher than in monolithic materials, these laminates exhibit good lightweight potentials due to the weight reducing aluminum in inner layers.

1. Introduction

In 1999, Saito et al. developed the accumulative roll bonding (ARB) process, which allows the severe plastic deformation of sheet materials, resulting in an ultrafine-grained microstructure. Each cold rolling pass with a thickness reduction by 50% results in an equivalent plastic strain of 0.8. By folding the sheets and repeating this procedure, an enormous strain is summed up, leading to an ultrafine-grained microstructure. The microstructural evolution and its influence on the mechanical properties during ARB processing has been discussed for various monolithic sheet materials. It was found that ARB processing can improve the strength and ductility simultaneously mainly due to Hall–Petch hardening. More recently, laminates combining different pure metals or alloys turned more and more into focus. Exemplarily, Jafarian et al. combined four different metals, Al, Cu, Zn, and Ni, to a multilayered laminate. They found out that the strength increases up to five ARB passes. However, necking of the Ni layer occurs after the first and of other material layers after the third and fourth ARB pass. The design of graded sheets with to special applications optimized properties is of strong interest. Apart from the main goal to obtain maximal mechanical properties after the repeated rolling passes, also the combination of mechanical and various functional properties is of great interest. The combination of two different types of aluminum alloys can improve the surface quality. Nanostructured Cu and Nb laminates offer an improved electrical conductivity or NiTi was already considered for biomedical...
In regard to mechanical properties, it is proven that laminated metal composites (LMCs) can also exceed the linear rule of mixture. So far mostly strength improvements have been discussed in the literature, although the influence on the cyclic fatigue behavior seems to be important, too. The different Young’s moduli in heterogeneous laminates, e.g., 110 GPa for titanium and 73 GPa for aluminum, cause complex stress states with stress jumps at the material interfaces. At an interface, both material layers possess the same strain (isostain model). Thus, applying Hooke’s law with two different Young’s moduli, the local stress differs and a stress jump occurs. For multilayered LMCs, elastic finite element (FE) analyses are necessary to describe the stress state in the laminate. The simulation shows the advantages of a LMC consisting of a stiffer core and a softer cladding. Here, the stress in the outer layer can be significantly reduced, which then lead to higher load-bearing capacities. In this work, two different layer architectures of Al–Ti laminates are tested by cyclic three-point bending. High bending fatigue lives and crack growth resistances are often essential in lightweight applications, for example, for the skin of an aircraft. The influence of ARB passes and the layer thicknesses on the three-point bending fatigue life of two ultrafine-grained Al/Ti laminates is investigated and compared to their ARB-processed monolithic counterparts. The test results are compared to the stress distribution in each layer of the LMC derived from an elastic FE analysis. For a final discussion, the nanohardness of each layer, measured by nanoindentation, and investigations of the crack path by scanning electron microscopy (SEM) and μ-CT are taken into account.

2. Experimental Section

2.1. Materials, Layer Architecture, and Processing

ARB processing was carried out with the aluminum alloy AA2024 (supplied by AMAG Austria Metall AG, Ranshofen/Austria) and titanium grade 1 (supplied by Harald Pihl, Täby/Sweden). The chemical compositions of both materials are listed in Table 1. Prior to the ARB process, the 2 mm-thick aluminum sheets were reduced to a thickness of 1 mm by five cold rolling passes. Subsequently, a solution heat treatment at a temperature of 493 °C for 1 h was carried out. The sheets were water-quenched and afterward stored at a temperature of −18 °C to avoid the formation of precipitates, prior to the ARB process. The titanium grade 1 was delivered in a solution annealed state and the sheet had a thickness of 1 mm.

The ARB process can be divided into four steps per pass. First, the sheets were cleaned with acetone to remove any grease and the surface was roughened by an electric driven wire brush in order to remove the oxide layer. Second, the sheets were stacked on top of each other. In the case of the first ARB pass, four sheets with a thickness of 1 mm were stacked according to different layer architectures, as shown in Figure 1a, to obtain a laminate with an outer aluminum or titanium shell. Third, the sheets were rolled with a thickness reduction of 50% by a four high rolling mill without any lubricant at room temperature (BW 300, Carl Wezel, Mühlacker, Germany). However, due to adiabatic and frictional heating the sheets are getting warm during the deformation. Due to the high deformation, the individual sheets are bonded together by cold-welding. The rolling as well as the used coordinate system is schematically shown in Figure 1b. To repeat the ARB passes (for N2 and N3 condition), the sheet is cut into two halves and again surface-treated and stacked. Because the ARB process was started in this work with four sheets, the number of individual layers in the resulting material is twice as high as in the original ARB process in accordance with Saito et al. The ARB process was carried out with the aluminum alloy AA2024 and Ti-1 layer. The indents had a lateral distance of 60 μm from each other.
2.4. FE Simulation of the Stress Distribution in LMCs Under Flexural Loading

Flexural loading of monolithic materials leads to a linear stress distribution over the thickness of the specimen. In the special case of a three-point bending, the stress distribution can be easily calculated by Equation (1) in the center of the beam, where \( z \) is the distance to the central fiber, \( F \) the applied force, \( l \) the support span, \( b \) the width, and \( h \) the thickness of the beam. Thus, the outer layer has to withstand the highest loads. In addition, the flexural stiffness \( E \) can be calculated with Equation (2) from the deflection \( \Delta z \) and load difference \( \Delta F \).

\[
\sigma_B = \frac{3Fl}{bh^2z} \quad \text{(1)}
\]

\[
E = \frac{\Delta F l^3}{4\Delta z bh^4} \quad \text{(2)}
\]

In laminates consisting of materials with different Young’s moduli, stress jumps occur at material interfaces because the same strain state (materials are fixed to each other) results in different stresses (Hooke’s law). The stress distribution becomes irregular. Thus, it is not possible to calculate the stress distribution analytically with Equations (1) and (2) and a finite element simulation is necessary.

The simulation was carried out using Abaqus FE software (Dassault Systèmes, Vélizy-Villacoublay Cedex, France). A 2D model of a beam with dimensions of 20 mm in length and 2 mm in thickness was applied (see Figure 3). The third dimension was included by the plane strain condition with a width of 9.5 mm. The whole beam was sectioned into 16 horizontal layers to realize the different laminates by assigning corresponding material properties to the individual layers. The material properties of aluminum were set to a Young’s modulus of 73 GPa and a Poisson’s ratio of 0.34 and, respectively, for titanium to 108 GPa and 0.32. For the FE mesh, 416,000 quadratic 8-node elements (CPE8) were used. Two nodes on the bottom of the beam were fixed, representing the support span, with a distance of 14.7 mm. In the middle node on the top side of the beam, a load of 177.5 N was applied. This particular load leads to a maximum stress of 100.0 MPa for monolithic materials. By comparison with the analytically calculated stress of 103.0 MPa (Equation (1)), a total deviation of 3% is obtained.

Apart from the stresses, the flexural stiffness of the laminates was calculated by the average displacement across the thickness of the beam by using Equation (2). The contour plots of the stresses in each layer architecture as well as the plot of the stresses across the thickness in the center of the beam are presented later in this work.

2.5. Three-Point Bending Fatigue Tests

Fatigue tests were performed on samples with 20 mm in length, 9.5 mm in width, and 2 mm in thickness, which was also applied in the FEM analysis. The samples were cut out of the ARB sheet with the long side parallel to the RD and subsequently ground.
down plane parallel using a grit size of 6 μm. Force controlled fatigue tests were carried out on a vibrophore testing machine (HFP 5100, Roell Amsler, Germany) in three-point bending mode with a support span of 14.7 mm. The forces for the monolithic samples as well as for the LMCs were calculated with the Equation (1) for the desired “nominal” stress amplitudes with a stress ratio \( R = \frac{\sigma_{\text{min}}}{\sigma_{\text{max}}} \) of 0.1. This analytical equation neglects the stress jumps at the interfaces between the different material layers, but allows a good comparability between the monolithic sheets and the LMCs. Subsequently, the real maximum stresses in the outer layers were calculated by FEM. Therefore, the factor by which the outer fiber stress of this LMC deviates from the mono material is multiplied to the nominal stress. The factors are the quotient out of the corresponding fiber stress of the laminate and the fiber stress of the monomaterial (100 MPa) (see Table 2).

The samples were fatigued at their individual resonance frequency of around 60 Hz. By monitoring the frequency, propagating cracks can be detected early by a decrease in frequency. The materials were tested at different stress amplitudes leading to fatigue lives in the range of 10^6–10^7 cycles. Runout tests, defined by more than 10^7 cycles, were stopped manually and afterward marked in the diagrams with an arrow.

2.6. Examination of Crack Propagation

The crack paths of several fatigued samples were investigated by using a SEM (Crossbeam 1540 EsB, Zeiss, Oberkochen, Germany) in the SE mode on cross sections. Additionally, the 3D crack network was analyzed in detail for the AlTiAl N3 laminate by using high-resolution X-ray microtomography (μ-CT) (SkyScan 1172, Bruker microCT, Kontich, Belgium). The material around the fatigue crack was cut off with a wire saw in a distance of 1.5–2 mm to reduce the X-ray specimen volume. The final dimensions of the AlTiAl sample were 3.5 × 9.5 × 2 mm³ (RD × TD × ND). The scans were performed with operating parameters of 80 kV and 100 μA and a wavelength of 0.024 nm in a rotation range of 180°. An Al/Cu filter was used to reduce the low energy photons below 30 keV. The machine was operating with an image pixel size of 3 μm. The recorded sinograms were reconstructed to 2D cross-sectional images using the reconstruction software (NRecon Client and Server 1.6.11 with GPU support, Bruker microCT, Kontich, Belgium). These images were assembled to 3D visual images using imaging software (Amira 6.3.0, FEI, Berlin, Germany) after segmenting with a global threshold on all layers.[20–24] Several images of cuts of the rolling—normal plane as well as of the rolling—transversal plane are presented by using the program DataViewer 1.5.2.4 (Bruker microCT, Kontich, Belgium).

Table 2. FEM calculation of the maximum tensile stresses for a load of 177.5 N. The highest stresses in each architecture are underlined.

| Material | Al or Ti | N1 | N2 | N3 |
|----------|----------|----|----|----|
| Max. stress in Al [MPa] | 93.4 | 84.8 | 82.4 | 33.6 | 57.1 | 69.5 | 100 |
| Max. stress in Ti [MPa]  | 68.4 | 91.1 | 103.8 | 105.7 | 114.4 | 117.5 |

3. Results

3.1. Microstructure and Bonding

Images of the microstructures of the AlTiAl and TiAlTi laminate after one to three ARB passes are shown in Figure 4. The stacking order of the sheets in the first ARB pass led to the fact that the two outer layers consist of one single layer, whereas all inner layers consist out of two layers of the same alloy. Figure 4a,b shows that the bonding of Al to Al, Ti to Ti, as well as Al to Ti was good and no inclusions or oxides were detected. Earlier investigations[16] on similar material showed that an interface affected zone (IAZ) with a thickness of 150 nm is developed between the titanium and aluminum layer. In the IAZ scanning transmission electron microscopy revealed higher dislocation densities and even smaller grains. Although the layers are quite straight after the first ARB pass, little waviness could be seen after two and three passes. Especially the outer layers which are in direct contact with the rolls exhibit variations of the thickness in RD. Exemplarily, the outer layer in the AlTiAl N3 laminate has a far thinner AA2024 top layer, whereas the bottom layer is quite near to the expected 125 μm.

Besides the layer shape, microstructural features could be seen in the BSE images. In the commercially pure titanium grade 1, small deformation structures appear. The microstructure of ARB processed titanium is discussed in detail in Terada et al.[25] The white particles in the AA2024 layers are typical precipitates of the S-γ-phase.[26]

3.2. Nanoindentation

The monolithic reference materials (Figure 5a) show an increase in nanohardness with increasing number of ARB passes: the nanohardness of monolithic AA2024 alloy increases almost linearly from 2.27 to 2.58 GPa and of Ti from 2.39 to 2.80 GPa. In both LMCs, the nanohardness of the outer layer material shows a higher increase compared to the reference material. Especially the nanohardness of the AA2024 in the AlTiAl laminate (Figure 5a) raises from 2.25 GPa (N1) to 2.93 GPa (N3). It exceeds the nanohardness of monolithic titanium after three ARB passes because the aluminum deforms in the composite more than the titanium due to its lower strength. This results in a higher work hardening. In contrast, the nanohardness of the Ti grade 1 in the TiAlTi laminate increases from 2.37 GPa (N1) to 2.79 GPa (N3). This nanohardness increase with the number of ARB passes is slightly lower than in the monolithic titanium. The layer material of the laminates, which is not in direct contact with the rolls, i.e., Ti for the AlTiAl and Al for the TiAlTi laminates, shows only a small increase after the second ARB pass and a small decrease after the third pass. This means that the nanohardness of the inner material does not reach the nanohardness of the monolithic materials. Especially the inner aluminum layer of the TiAlTi laminate shows a decrease in nanohardness after the third ARB pass. This decrease in material hardness may be, on the one hand, due to different positions of the indent arrays. In the case of N3 laminates, the indent array was put around the third interface relatively to the last bonding interface (compare Figure 2). In contrast, the indent array for the N2 laminate is positioned at the second interface.
and thus it is much closer to the last bonding plane. It is well known for ARB that the local shear strain is maximal at the surface and might fluctuate strongly across the thickness, depending on the number of ARB passes. On the other hand, recovery and/or local dynamic recrystallization can also come into play after the high plastic straining by three ARB passes. In this context, it has to be reminded that although the ARB process has been performed at RT, the temperature of the sheets increases significantly due to adiabatic and frictional heating during ARB processing. Measurements of the sheet temperature directly after rolling revealed a maximum temperature of about 80 °C.

### 3.3. Simulated Flexural Stresses

The maximum stress in the monolithic material reaches 100 MPa at a load of 177.5 N, allowing simple calculations of the deviations to the stresses in the LMCs. In Figure 6, the stresses for the whole beams are shown in means of contour plots and diagrams of stresses across the thickness in the center of the beams. As expected, the simulation of a monolithic material reveals a linear increase of the stress across the thickness, independent of the elastic properties of the material. In contrast, the LMCs exhibit pronounced stress jumps at the material interfaces, which can be clearly seen in the stress thickness diagrams. In the case of AlTiAl laminates, the stress in the outer fiber is lowered compared to monolithic material due to a local redistribution of the load from Al to Ti, whereas the maximum flexural stress in the outer fiber increases in the TiAlTi laminate compared to the monolithic material. In the AlTiAl N2 and N3, the maximum flexural stress in the inner Ti layer exceeds the stresses of the outer Al. The values of the maximum stress in the Al and Ti layer are given in Table 2 calculated for each laminate. Maximum values are observed for the N3 laminates. The stress in the outer fiber of an AlTiAl N3 laminate is reduced due to stress redistribution to the inner Ti layers by 18% in

![Figure 4. SEM images of the a,c) AlTiAl and b,d,f) TiAlTi laminates in the RD–ND plane after one to three ARB passes. The Ti layers are darker than the AA2024 layers due to the mass contrast of the BSE imaging.](image-url)
comparison to the monolithic case. However, this lower stress leads in that case also to a huge increase of the stress in the Ti layer below the outer Al layer. The Ti layer must withstand 3.8% higher stress compared to the monolithic case at same load.

The flexural stiffness (Equation (2)) of the laminates was also calculated with the deflection and force data of the simulation (see Figure 7). The beams of the monomaterial behave around 4% stiffer than the input elastic constants for Al and Ti (73, 108 GPa). The flexural stiffness of the laminates lies between the monolithic Al and Ti and convergences with increasing number of alternating material layers to the average of the elastic moduli of the reference materials. Depending on the architecture of the laminates, the flexural modulus for the N1 laminates deviates strongly from the expected linear value of mixture, calculated to 94.4 GPa. However, with increased number of ARB passes and thus decreasing layer thickness, the flexural moduli are approaching the expected mean value. Generally, the TiAlTi laminate reveals a higher bending stiffness for all ARB passes compared to the AlTiAl laminate because of the predominating stiffness of the outer layer.

Figure 5. Nanohardness of a) the monolithic reference materials, b) AlTiAl and c) TiAlTi laminates with separated nanohardness values for each layer material.

Figure 6. Contour plots of the stress distribution of laminates with an outer aluminum layer (left) and an outer titanium layer (right) after different ARB passes in comparison to monolithic material. The stress distribution in monolithic materials is independent of the Young’s modulus, whereas laminates exhibit stress jumps due to different Young’s modulus. In the upper part of the beams, artifacts of the point load appear.
3.4. Flexural Fatigue Properties

Three-point bend fatigue tests were performed for the laminates and monolithic materials with one and three ARB passes. The results are shown in a Woehler S–N diagram (see Figure 8). For the laminates, a “nominal maximum bending stress amplitude” has been calculated by using the analytical formula for monolithic beams (Equation (1)). The results of Figure 8a,b allow a good comparability between all different states because the complex stress states are neglected in the laminates. In the diagrams below, Figure 8c,d, the simulated maximum stress of the outer fiber, derived from the abovementioned finite element analysis (Section 3.3), was considered.

Titanium grade 1 shows for both tested conditions with different ARB passes, the best fatigue behavior compared to the other tested materials. It must be pointed out that due to the limited number of tests performed up to now, the absolute position of the fatigue life curves will vary; however, the general trends became clear. After one ARB pass a fatigue strength of 300 MPa is obtained at $2 \times 10^6$ fatigue cycles, while after three ARB passes 330 MPa is reached. An improved fatigue life is also observed with an increasing number of ARB passes in the LCF regime. In contrast, the aluminum alloy AA2024 shows a fatigue strength of 200 MPa for both the N1 and N3 conditions. The LCF life is also significantly affected by the number of ARB passes. Compared to Ti grade 1, the transition from the LCF to HCF regime takes place at a lower number of cycles.

The nominal endurance fatigue strength at $2 \times 10^6$ cycles of the AltaiAl laminate (see Figure 8a,b) accounts to 225 MPa for the N1 and 205 MPa for the N3 condition. Contrary to this decrease of fatigue life in the HCF regime, the LCF life is slightly improved performing three ARB passes. Compared to monolithic AA2024, the AltaiAl laminate exhibits in the LCF region for both tested ARB passes higher fatigue lives. Contrary, the fatigue life in the HCF regime is after one ARB pass around 25 MPa higher and after three ARB passes similar to that of AA2024 after three ARB passes. The AltaiAl laminate demonstrates, especially in the LCF region, a significantly higher fatigue life compared to the AltaiAl LMC. The endurance fatigue strength at $2 \times 10^6$ cycles lies at around 245 MPa for both performed ARB passes. It is around 20–40 MPa higher than that of the AltaiAl laminates. Compared to monolithic titanium, the difference in the LCF region is after one ARB pass slightly lower than after three ARB passes. The gap between the Ti and AltaiAl S–N plots in the HCF is also larger after the third ARB pass. It also seems to be prevalent that for both laminates the slopes of the fatigue life curves in the HCF regime seem to be steeper for the N3 condition than for the N1 laminates. However, further fatigue tests have to be performed in order to prove this behavior.

In Figure 8c,d, same S–N results are shown on the basis of the effective stress in the outer fiber of the laminates, calculated by the FE analysis. This can be done because the fatigue cracks initiate at the surface of the materials, where the highest stresses appear. Moreover, an elastic simulation is sufficient because the stresses in the fatigue experiments are below the yield strength. As shown in the simulation, the effective stress in the outer fiber of the AltaiAl laminate is 6.6% lower than calculated with the analytic equation for monolithic materials (Equation (1)). Thus, the AltaiAl shows after one ARB pass nearly the same S–N curve as the monolithic AA2024 (see Figure 8c). As the outer Al layer gets very thin for the N3 laminate, the inner Ti carries more load and lowers the outer stress by around 18%. Thus, the endurance effective fatigue strength of the outer Al layer is lowered to around 170 MPa, much below the monolithic material. The fatigue lives in the LCF regime are nearly equal to that of monolithic AA2024. The AltaiAl laminate shows compared to the monolithic simulation an increase of the outer stress of 6% for the N1% and 18% for the N3 condition. Regarding these results, the AltaiAl N1 fatigue lives in the LCF regime become rather closer to that of monolithic Ti. This small gap gets even closer after the third ARB pass, where the fatigue lives are almost identical. The stress amplitudes in the HCF regime come also closer to that of monolithic Ti.

3.5. Crack Propagation

Crack propagation plays an important role in fatigue and contributes to different amounts to the total fatigue life. In the case of bending fatigue, cracks initiate at the outer fiber, where the highest stress values occur. The number of cycles needed to initiate a macroscopic crack $N_f$ obtained from frequency drop analysis, as well as the number of cycles to fracture is plotted in Figure 9.

In the LCF regime, crack initiation in AltaiAl N1 occurs at around 69% of the total number of cycles to failure $N_f$ (see Figure 9a). This is significantly earlier than in the case of a TiAI laminate and monolithic counterparts were $N_f$ amounts to 90%. The $N_f/N_i$ ratios of the laminates with three ARB passes become similar to each other (0.8). In contrast, the monolithic materials show an unchanged higher $N_f/N_i$ ratio of 92%. Compared to the monolithic materials, crack propagation in the LMCs is retarded. In particular, the Al to Ti interlayer in the AltaiAl N1 laminate seems to impede the crack propagation most effective. In the HCF region, where crack initiation mainly governs the fatigue life, $N_f/N_i$ ratios of 97–99% were determined for both laminates and monolithic materials. Thus, crack propagation becomes less important for the total fatigue life.

SEM was used to evaluate the influence of the different layer architectures and interfaces on the crack paths. Figure 10 shows...
Figure 8. Woehler S–N plots of the ARB materials a) with 1 and b) with 3 ARB passes. Due to a limited number of data points, some of the S–N curves are only indicated by dotted lines. c,d) The difference between the nominal stress amplitude and the effective stress amplitude in each outer fiber calculated by FE analysis (dashed lines).

Figure 9. Stress amplitude versus the number of cycles for crack initiation N_i and failure N_f for a) N1 and b) N3 conditions. The numbers next to the curves are the average ratio N_i/N_f in the LCF regime.
exemplarily typical crack paths for the laminates and the monolithic counterparts after three ARB passes (16 layers) of samples fatigued in the LCF region. Cracks in the monolithic materials proceed mostly to the central layer, which is the last cold-welded and thus weakest interface. In case of monolithic AA2024 (Figure 10), the crack is often diverted, which probably might be caused by coarse precipitations. The monolithic Ti shows a widely opened main crack accompanied by smaller side cracks perpendicular to the driving force. The latter behavior might, to some extent, caused by the locally weaker bonding interfaces. In the AlTiAl laminate (Figure 10c), multiple cracks initiate at the outer Al layer. Some of them are deflected or stopped at the first Al–Ti interface (arrows). The main crack propagated further to the second Al–Ti interface, where it was diverted. Furthermore, every AlTiAl specimen delaminated along the interface between the two inner Al layers due to a lower bonding strength of the last ARB bonding plane. In the TiAlTi laminate, the wide opened crack was diverted at the first Al–Ti interface. Only a small crack propagates straight ahead into the titanium layer and stops there. Both typical crack networks of the laminates show crack stopping at the transition to Ti. Thus, the AA2024 seems to be the weaker component in the LMCs. Moreover, in AlTiAl, the Ti layer shows pronounced plastic zones ahead of the stopped cracks, which is rather similar to the behavior observed for the monolithic Ti. In contrast, in the TiAlTi laminate, no plastic zones are visible around the crack (see Figure 10d). A reason for that could be the different stress levels, which the two laminates can resist at similar fatigue lifes. Exemplarily, the stress of the AlTiAl sample (Figure 9c) was 75 MPa lower than in the TiAlTi laminate (Figure 9d). Additionally, the effective stresses are reduced in the Ti layers of the AlTiAl laminate due to stress redistribution (compare FEM results in Figure 6). Further, the stronger shear deformation of the surface material, caused by the friction between the roll and the sheet surface,[28] leads to strengthening. This results in a strength gradient from the surface to the middle of the laminate.[27]

Additionally, a 3D recording of the crack network was provided by micro-CT measurements (Figure 11). Two different kinds of sections were stacked to analyze these data. Nine cuts were taken parallel to the RD and ND with about 1 mm between each other (Figure 11a-i). The crack propagation in these views can be compared to the side view of the SEM recordings from Figure 10. In addition, four cuts of the RD–TD plane were created. These cuts were taken from the four lower Ti layers because the cracks were only visible in the Ti material and propagated only till the middle layer. Thus, it becomes possible to see interruptions or bifurcations of the crack path along the width of the sample. Especially in Figure 11j, three independent cracks can be distinguished in the first Ti layer (arrow 1). The two smaller cracks were stopped within the Ti layer, as they do not appear in the next image of the outer Ti layer (Figure 11k). The larger crack propagates to the next layers. As shown in Figure 11b,c, the crack propagates in the following Al layer to the right side (2), whereas the crack propagates in Figure 11d-g to the left side (3). The main crack branches at the second Al–Ti interface or in the Al layer. A high degree of bifurcation can be seen in Figure 11e-g. The crack propagates here to the middle Al–Al interface (4) and delaminates within the last bonding plane. The μ-CT proves that the crack branches in the laminates in various directions. The pronounced branched crack paths in the
LMC are in good accordance with the analysis of the \( N_i/N_f \) ratios, described above. Consequently, the number of cycles for propagation is enlarged in the LMCs. Based on the \( \mu \)-CT results, crack bifurcation and zigzag-type crack propagation are the reason for the extended crack propagation regime.

4. Discussion

The microstructural analysis showed the good joining between the AA2024 and Ti alloy because no significant inclusions or defects are observed at the boundary. This is a prerequisite for good mechanical properties and the applicability under bending stresses. Thus, the S–N curves of TiAlTi exceed the linear rule of mixture in the LCF region and are almost equal to the linear rule of mixture in the HCF regime. If considering the effective stresses in these LMCs calculated by FE analysis, the stress amplitudes of the TiAlTi laminate almost reach those of monolithic Ti. In contrast, the fatigue strength of AlTiAl in the HCF regime could not reach the rule of mixture, although the effective stress in the outer fiber of the AlTiAl laminates is lowered by 6.6% (N1) and 18% (N3). After one ARB pass, the effective stress in the outer fiber is identical to the fatigue strength of monolithic AA2024, but after three ARB passes, the effective fatigue strength in the outer fiber is around 40 MPa lower. Various reasons can be found for the fatigue life difference between the LMCs under three-point bending: the rolling process has an enormous influence on the strength of the materials.
Lee et al.\textsuperscript{[27]} reported that the high friction during rolling leads to high shear deformation in near-surface regions. In the case of the AlTiAl laminates, the softer AA2024 is in direct contact with the roll in every ARB pass and lies on a stiffer and stronger titanium. This leads to a more pronounced shear deformation and compression of the material, resulting in the strengthening of the AA2024 layer. This is shown by nanoindentation, where a rise from 2.25 GPa (N1) to 2.93 GPa (N3) was detected. For comparison, in the case of the TiAlTi laminate, the rise in nanohardness in the Ti layer is only from 2.37 GPa (N1) to 2.79 GPa (N3). Also, in comparison to monolithic AA2024 (2.58 GPa, N3), the AA2024 layer in AlTiAl exhibits a higher nanohardness. A further reason for the different hardness development is the faster grain refinement in the AA2024 compared to the pure titanium, due to its alloying elements. This was exemplarily shown for bimodal microstructures of Al99.5/Al99.9 laminates, where the highly alloyed layers exhibited smaller grains and thus higher hardness.\textsuperscript{[28]} The high hardness in AlTiAl is also combined with a loss of ductility in the outer AA2024 layer, which then might decrease the crack resistance. The rather brittle behavior of AA2024 after several ARB passes was already shown in a precedent study,\textsuperscript{[16]} as well as in\textsuperscript{[6]}\textsuperscript{[6]}. Moreover, the different crack paths in the LMCs (compare SEM images of Figure 10) support this assumption. At a first glance, titanium seems to behave more ductile in the AlTiAl laminate as observed at the plastic dimple structures around small cracks. However, the crack is diverted at the hard Ti layer in the TiAlTi laminate, without visible plastic zones. Also, the microstructure of AA2024 with its brittle and coarse precipitates (S’/Q-phase) and intermetallic phases\textsuperscript{[29]} influence the fatigue life of AlTiAl laminates. Together with the abovementioned low ductility, cracks might be easily initiated or accelerated at precipitates. Contrary to the crack arrest at the transition from the Al to the Ti layer is the stress distribution in the laminates, which was calculated by FEM. The results suggest that crack initiation and propagation would be accelerated due to up to 17 % higher stresses in the Ti layer (compare FEM results in Figure 6).

The different stress distribution as well as the different strength and hardening of the two alloys cause different crack propagation behaviors of the laminates, which were examined by SEM and μ-CT. In general, it can be concluded that an initiated crack at the surface of a TiAlTi laminate is driven by huge normal stresses and propagates perpendicular to the surface. The subjacent AA2024 layer is not capable to stop the propagating crack, whereas initiated cracks in the outer AA2024 layer of AlTiAl can be stopped by the subjacent Ti layer due to its strong and ductile behavior. This is also verified by the ratio of $N_i/N_f$ (see Figure 9) because the ratio highlights the dominant crack propagation regime in the AlTiAl N1 LMCs. Thus, crack propagation is strongly influenced by the stacking order of the sheet layers. It was shown that a crack coming from the AA2024 layer is impeded at the boundary to titanium. In contrast, at a Ti–Al interface, where the crack propagates from the stronger to the weaker material, no bifurcation at the boundary occurs. This crack propagation behavior is schematically illustrated in Figure 12. Similar behavior was shown for an iron–steel transition in Pippan et al.,\textsuperscript{[90]} where crack bifurcation occurred in front of the plastically stronger steel.

5. Conclusion

The present work examined the influence of two different layer architectures of AA2024 and pure titanium on the stress distribution and its effect on the three-point bending fatigue strength. Different Young’s moduli of layers in laminates lead to the redistribution of stresses under bending loads. Thus, the maximum surface stress depends strongly on the layer architecture. FE analysis revealed that the stacking order reduces the outer fiber stress by 18% in the case of an AlTiAl N3 laminate and increases in the TiAlTi laminate compared to the monolithic reference material. Although TiAlTi laminates have to withstand higher outer fiber stresses, fatigue lives are higher compared to the TiAlTi laminates due to lower fatigue properties of AA2024. This effect becomes even more pronounced after the third ARB pass due to different nanohardness evolutions in the LMCs. Consequently, the $S$–$N$ curve of TiAlTi exceeds the rule of mixture and thus highlights the potential of optimized LMCs exhibiting a good lightweight potential. Further optimization of layer architectures, the use of

![Figure 12. Schematic illustration of the crack propagation at transitions from the weaker AA2024 to the stronger Ti and vice versa. At the transition from the weaker to stronger material crack bifurcation occurs. The crack propagation behavior is in good accordance with the publication of Pippan et al.\textsuperscript{[90]}](image-url)
other materials, or the use of multimaterial laminates will lead to a further improvement of the strength of LMCs, especially under flexural loading conditions, where a nonlinear stress distribution can help to increase the load-bearing capacity.

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**Conflict of Interest**

The authors declare no conflict of interest.

**Data Availability Statement**

The data that support the findings of this study are available from the corresponding author upon reasonable request.

**Keywords**

accumulative roll bonding, fatigue, FEM, laminated metal composites, layer architecture

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